METALLURGIA

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INCORPORATING "THE METALLURGICAL ENGINEER."

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The new Avro-Manchester twin-engined aircraft now in service with the Bomber Command of the R.A.F.

Aluminium and Aluminium Alloys in Peace and War

By William Ashcroft

About forty years ago the demand for aluminium was negligible, but continuous improvements in the production and fabrication, and continuous research to improve its alloys, have contibuted to it becoming one of the most versatile of metals. For the duration of the war it has practically disappeared from the materials used in manufacture of products for non-military purposes. This article briefly reviews the trend of development and summarises some of its uses in modern warfare.

IN the past fifty years aluminium has advanced from what was once regarded as a laboratory curiosity to a position of vital importance among the base metals. It is generally admitted that the introduction of a new metal into the service of mankind is an operation of extreme difficulty, and it speaks well for the properties of this material that it should have made such progress in a relatively short time. To-day, it is difficult to appreciate that in 1905 world production of aluminium had reached only 10,000 tons per annum: in about thirty years afterwards, however, consumption was estimated at 400,000 tons per annum, and it is probable that, in present world conditions, production is approaching a million tons per annum.

Continual progress in improving the economic utility of aluminium has been the important factor in the marked advance of the world industry. Some special circumstances, such as the rise of nationalism, have also influenced this progress, especially during more recent years. The field of application of aluminium, up to the outbreak of present hostilities, showed constant expansion as a result of its increasing usefulness, which has been convincingly demonstrated. The discovery that aluminium was especially suitable for cooking utensils assisted the early development and application of the metal, but gradually it began to be introduced in engineering. Aluminium is too weak for many engineering purposes, and aluminium alloys began to appear; but at first the manufacture of these alloys progressed slowly, due largely to the reluctance of engineers to adopt new materials in established practice. Even immediately before the present war opposition and prejudice existed among some of the old-established industries against the use of aluminium: it is not surprising, therefore, that difficulties in marketing this new material and its alloys were encountered in its early days.

The application of aluminium and its alloys received great impetus from the automobile industry, and it was fortunate this new industry was born when the new materials were struggling for recognition. This new industry, starting with a more open mind and a need for light weight and other properties possessed by aluminium alloys, appreciated the possibilities of the new material and became a substantial consumer. It was the coming of the aircraft industry, however, and the need for light-weight construction of both aircraft and engine, that fostered the development in quality of aluminium alloys and greatly increased their application in recent years.

The enhanced utility of aluminium was largely due to a number of discoveries and outstanding technical achievements. The relatively low strength of aluminium undoubtedly retarded its application for many purposes, and while some progress was made in increasing strength by using copper as an alloying element, the discovery by Wilm, in 1910, that an aluminium alloy, containing 4% copper, 0·5% magnesium, 0·5% manganese, and the remainder aluminium, could be hardened, proved to be

One of a number of the latest type Spitfire fighter planes which has been presented by the Bermuda War Fund. These planes are powered by Rolls Royce Merlin engines, giving a speed of approximately 400 miles per hour.

a development of major importance and gave an impetus to research which rapidly intensified. From this discovery the world-famous wrought alloy, Duralumin, was developed. Later, during the period of the last great war, the late Dr. Rosenhain and his collaborators at the National Physical Research Laboratory developed the well-known Y alloy for high-temperature applications, and especially aero-engine pistons and cylinder heads. For some time the application of these alloys developed, but the increasing demands for light alloys of higher strength were insistent, and early among the achievements in this direction were the series of R.R. alloys developed by Hall and Bradbury, of Rolls-Royce, Ltd.

The high physical properties of all these alloys are developed by heat-treatment, which may comprise solution treatment or solution treatment and subsequent precipitation treatment. Among the early achievements was the development of an alloy possessing a resistance to corrosion corresponding with pure aluminium, but of much greater strength. This aluminium alloy, containing 3·5% magnesium and 0·5% manganese, known as "Birmabright," and the forerunner of a series of corrosion-resistant alloys, in certain wrought forms has a tensile strength as high as 25 tons per sq. in. This tensile strength is developed solely by cold-working: it is not necessary, therefore, to subject completed work to heat-treatment to develop the properties desired. The aluminium-silicon alloy was another development which contributed to the increased utility of aluminium. The excellent casting qualities of these alloys, combined with their corrosion resistance, make them especially suited for many purposes.

It is not possible to mention all the alloys that have been developed as a result of the intensive research work on aluminium and its alloys, but brief reference should be made to important work in more recent years in connection with the production of high-purity aluminium. All these developments have contributed to the enhanced utility of aluminium and to the expansion of industry. Parallel with these developments, of course, are the great strides made in manufacturing technique. Sand-, die- and ingot-casting processes have become highly developed, while on the wrought side, pressing, extrusion, tube-drawing and rolling of sheet and strip have been added to the older forging and stamping processes of working high-strength aluminium alloys. Great progress has been made in heat-treatment operations and in equipment for these operations; also in equipment for the production of wrought manufactures, large shapes, methods of fabricating, welding, structural design, foil, powder and surface finishes.

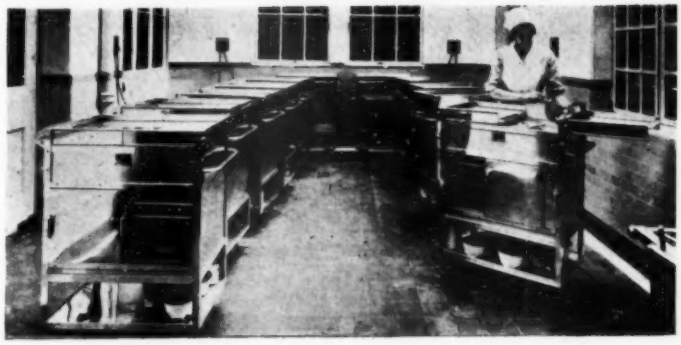
As a result of technological progress in these and other directions many new products have been developed, and the market for aluminium, at the beginning of the present war, had greatly broadened. Thus, the use of aluminium had been extended to purposes for which it was formerly uneconomical in competition with other metals and materials, and its usefulness has been convincingly demonstrated as time has passed.

All the aluminium-producing countries contributed and benefited from the results of research and technological progress, but there are indications that the institution of the totalitarian systems is responsible for the artificial increase in production, an increase closely connected with the production of armaments. The first noticeable increase took place in Germany in 1934, with the justification that aluminium being a German metal had to replace the then imported metals, principally copper and tin, and to some extent lead. These three metals and their alloys were on the list of the "Sparmetalle"—i.e., metals to be spared. In this country, or in the United States, they would be called "strategic" metals, but this designation would have revealed the real intention of the greatly enlarged aluminium production in Germany.

During the period 1934 to 1939, prior to the present war, the so-called peaceful activities of Germany were indicated by the application of aluminium and alluminium alloys in the production of field-kitchens, water-bottles, reflectors, searchlight carriages, radio equipment, gun-wheels, pontoon bridges, truck equipment, bomb fuses, cartridge cases, magnetic, acoustic and moored mines, paint, cars, buckles, food containers, and many other applications, especially in ship equipment and aircraft construction, and also as a substitute for other non-ferrous metals. As a result of this evolution all the aluminium produced in Germany could not be made available for direct military use, since a substantial amount had to be kept to replace copper, brass, lead and tin, but tremendous stores of finished aluminium products had been built up for war purposes.

In Great Britain, and later in the United States, it has been possible to divert practically the whole of the aluminium production of these countries to work of military importance.

This change was only made as each country moved from peace-time industrial activity to belligerent conditions. How great the change is not readily appreciated, but when it is realised that the versatility and diversified uses of aluminium and its alloys in thirty major industries are represented by some 2,500 applications it may be possible to visualise the effect of such a change. These applications vary appreciably from one country to another, depending on the degree of industrialisation and the standard of living, but some guide to an estimate of world consumpiton of aluminium in various fields is given in Table I.



Electrically heated hospital tables in Birmabright.

Aluminium, in its relatively short history, has won a predominating position in aircraft engineering. The importance of its alloys for aircraft structures and engines is due to the combination of four qualities—low specific gravity, strength, corrosion resistance, and adaptability to various manufacturing methods. Apart from specific gravity, these qualities have been developed to an extent that exceeds the dreams of the early workers of this metal without appreciably affecting its low specific gravity.

TABLE I.
ESTIMATED WORLD ALUMINIUM CONSUMPTION OVER TEN-YEAR PERIOD.

	%
Aircraft, automobile, railway, marine industries	30
Food industry, cooking utensils	15
Large and small machinery	15
Electric industry: conductors, busbars, motors	10
Chemical industry	3
Deoxidising steel and heavy alloys	7
Powder, foil, seals	4
Non-recoverable scrap	3
Miscellaneous	13

The gross weight of an aeroplane, according to Haenni,* is made up of the weight of the airframe, accessories, engine, propeller and load. In a four-engine 25-ton bomber the structural weight may amount to 35%, leaving about 20% for the power-plant weight and about 45% for the variable load. The approximate distribution of materials is shown in Table II, which gives some idea of the position occupied by aluminium relative to other materials. It will be appreciated that this relationship varies with the type of aircraft, and also to some extent with the design, but the amount of aluminium consumed in the construction of various types approximate to that given in Table III.

TABLE II.
WEIGHT ANALYSIS OF HEAVY BOMBER IN POUNDS.

Metal.	Form.	Wing.	Fuselage.	Tail. Surface.	Landing Gear.	Controls.	Total.
Steel	Low alloy	300	200	20	1,000	80	1,600
	Stainless	500	50	—	—	50	600
Magnesium	Castings	—	—	—	—	50	50
Aluminium	Sheet	7,000	2,100	800	100	500	10,500
	Extrusions	150	100	50	25	175	500
	Drawn parts	950	400	100	—	50	1,500
	Castings	—	—	—	—	100	100
	Forgings	150	50	25	100	25	400
	Rivets	250	100	50	—	50	450
Others	All	20	20	20	40	100	200
Total for Structure		—	—	—	—	—	15,900

TABLE III.
APPROXIMATE AMOUNT OF ALUMINIUM IN WAR PLANES.

	Lb. Net.		Lb. Gross.
Fighter	5,000	..	6,000
Dive bomber	10,000	..	12,000
Medium two-engine bomber	18,000	..	21,000
Heavy four-engine bomber	30,000	..	34,000

Some parts for both the structure and the engine are usually from non-treatable aluminium alloys, but practically all the stressed parts are made of heat-treatable alloys, and sheet is used wherever possible. The parts usually made of non-treatable aluminium in sheet form are cowlings, covers, fairings, leading edge, flooring, tank, unstressed fittings, and many miscellaneous parts. In some cases magnesium alloys are substituted for aluminium for these parts. In tube form the parts include fuel and oil lines, air-speed indicator tubes, conduits and other miscellaneous parts.

In engines it is more difficult to differentiate between forged and cast parts because of the varying technique employed, the type of engine, whether air- or water-cooled, and upon the stress imposed upon the part. Thus, the crankcase and cylinder heads of water-cooled and small sized radial engines are usually cast; the cylinder head of high-powered air-cooled engines may be cast, but in Britain are usually forged. The piston rod section and covers are sometimes cast for small air-cooled engines, but forged for high-powered ones. The parts which are nearly always cast are the blower-section, the rear-end section, carburettors, supercharging casing, pump housing and intake section. The parts which are mostly forged in radial engines are the three main crankcases, supercharger impeller blades, engine support, landing gear, fitting, etc. Some parts may either be made of aluminium or some other material. This applies to propellers, bearings, and even the crankcase of radial engines. Aircraft parts for which the magnesium alloys are sometimes favoured are landing wheels, oil pump, supercharger, front and rear sections, instrument housing, control parts, distributors, accessories, plates, covers, etc.

Although only one aspect of the part played by aluminium in modern warfare has been briefly summarised, it is the most important application. Brief consideration has been given to the considerable progress in the development of stronger alloys and in the fabricating technique applied to them. This progress has been of an international character, and the alloys used in German, Italian or Japanese planes are similar to those of corresponding types built in Russia, United States or Britain. It will be appreciated, therefore, that whatever the development achieved in this country it does not permit a feeling of complacency, as there seems to be still some margin left for improving aluminium alloys with respect to their mechanical and chemical properties and their adaptability to mass production.

Still higher strength means lighter construction—i.e., increased speed or safer construction, or a higher operating ceiling. There is a need for light alloys with even higher strength for propellers, with even greater hardness at high temperature for aero-engine pistons, with even superior casting properties for engine crankcases, with even better machining properties, and with even higher fatigue characteristics for aircraft covering. The nations hostile to us appreciate the significance of such progress, and their efforts to achieve the development of superior light alloys and improved manufacturing processes is indicated by their recent patents and publications.

There is ample reason to feel a certain pride in the aluminium alloys now being supplied to aircraft manufacturers in the British Empire, but it is also significant that the aluminium industry is alive to the fact that these materials may not meet all requirements a year hence, and is vigilant in observing and applying every advance in the science of metallurgy to enhance the effectiveness of laboratory and manufacturing activities.

* P. M. Haenni. "Light Alloys in Modern Warfare." Paper before Montreal Chapter, American Society of Metals. *Canadian Metals*, 1942, 5, pp. 36–42.

The Carbide and Nitride Particles in Titanium Steels

IT is known that titanium steels usually contain not only the constituents of the Fe–Ti–C system, but also small particles of variable colour, which are due to the presence of nitrogen. Great uncertainty has existed as to the nature of these particles, which have been described by different investigators as "nitride," "cyanide," and "cyano-nitride." The object of a recent investigation, described by W. Hume-Rothery, G. V. Raynor, and A. T. Little,* was to determine the nature of these particles and to show the relations existing between titanium carbide TiC, and titanium nitride, TiN. For this purpose a series of titanium steels of commercial purity, containing from 0 to 6% of titanium and from 0·1 to 1·0% carbon, was made available by the Research Department, Woolwich. The steels were prepared in a spark-gap high-frequency furnace, and the ingots were forged to $\frac{5}{8}$ in. dia. bars, which were supplied in the furnace-cooled condition. They were examined microscopically in this state, and also after annealing for a varying number of hours at 700°, 800°, and 1,000° C. respectively, followed by quenching in ice and salt. Some of the bars were then submitted to anodic attack in 2N hydrochloric acid; the resulting residues were examined by X-ray and chemical methods, and the results obtained have thrown considerable light on the carbide-nitride relations in this series of steels.

The preliminary microscopic examination showed clearly that a proportion of the titanium was present in the form of impurity particles or clusters, and that this proportion was variable, and apparently quite uncontrolled. For this reason it was not possible to construct accurate equilibrium diagrams, since it is not always possible to determine the titanium present in the impurity particles and clusters, as distinct from that in the steel as a whole. The structures found in the furnace-cooled alloys, and in those quenched from 700°, 800°, and 1,000° C. respectively, are given in four diagrams, but it is emphasised that the diagrams are very approximate, since in many cases a considerable proportion of the titanium, as well as some of the carbon, is present in the impurity particles and clusters. These diagrams merely connect the structures of the particular batch of steels examined with the compositions of the steels as a whole.

Special attention was given to the nature of the impurity constituents which are usually referred to as particles of titanium nitride. Systematic examination of these impurity particles showed that increasing carbon content produced a regular series of changing colours of these particles in the direction yellow → orange → brown → grey. Further duplex or rimmed particles frequently existed in which the centre was yellow or orange and the outside orange, brown or grey; in all cases the inside of the duplex particle was nearer to the yellow end of the above series of colours than the outside. In the steels of high titanium and carbon contents, many of the particles of the impurity clusters were either entirely grey or had grey rims, which were almost indistinguishable from the titanium carbide particles present as a genuine constituent of the Fe–Ti–C system.

The microscopic investigations shows clearly that a whole range of impurity particles exists, and the authors suggest two interpretations. Either (*a*) a series of solid solutions may be formed by TiN and TiC, and the duplex particles may be the result of severe coring, or (*b*) compounds which are essentially nitrides may solidify first and then undergo peritectic reactions with the melt to produce carbide. The work described in succeeding parts of their report suggests that the former alternative is the more probable.

Residues were extracted from the steels by electrolytic attack in hydrochloric acid, and were submitted to chemical analysis and X-ray examination. The chemical tests showed that the yellow, orange and brown particles were attacked by chlorine at 300° C. with the liberation of free carbon and formation of volatile titanium tetra-chloride, whereas titanium carbide is unattacked and titanium nitride is attacked with the formation of volatile titanium chloride. The analyses suggested that in the orange and brown crystals some of the nitrogen of titanium nitride was replaced by carbon.

The X-ray investigation of the residues showed that the yellow, orange and brown crystals gave rise to diffuse diffraction lines corresponding to face-centred cubic structures with lattice spacings between those of TiC and TiN. The diffuseness of the lines varied greatly in different residues.

The main conclusions drawn by the authors from this work are summarised as follows:—

1. The microscopic X-ray and analytical results are all in agreement, and suggest that the impurity particles in the titanium steels involve a series of solid solutions between titanium carbide and nitride.

2. The general effect of increasing carbon content is to change the colour of the impurity particles in the direction yellow → orange → brown → grey, and duplex particles are frequently observed. Increasing titanium content causes the same change, although to a less marked extent. The distribution of colour is, however, very variable, and different parts of the same bar may give different proportions of the various colours.

3. The X-ray results show that the lattice spacings of the yellow and brown particles lie between those of TiC and TiN, and suggest that a wide range of compositions exists in the steel containing 5·8% titanium and 0·58% of carbon.

4. The chemical analysis shows that part of the residue from anodic attack of hydrochloric acid is decomposed by chlorine with the liberation of free carbon and the formation of volatile TiCl. The numerical values suggest that replacement of nitrogen in TiN by carbon is taking place in the yellow, orange or brown particles.

* Iron and Steel Institute. March, 1942. Advance copy.

Beryllium Copper

DESIGNERS and metallurgists have been familiar for some years with the unique properties of beryllium copper. The combination in this alloy of exceptionally high tensile and fatigue strengths, great hardness, good electrical conductivity and resistance to wear and corrosion has established its usefulness where such properties are demanded for current carrying springs and other parts of electrical equipment, for instrument springs, diaphragms and capsules, and for a variety of other components in precision equipment.

An improved ternary alloy of this type, developed from the original beryllium copper by the addition of a small percentage of cobalt, is marketed under the name Mallory 73 Beryllium Copper. The addition of the cobalt result in somewhat improved and more uniform physical properties, and springs or other parts of more complicated shape may be fabricated from this alloy in the soft state. Subsequently, a short heat-treatment develops its remarkable physical properties—a tensile strength up to 90 tons per sq. in. with a proportional limit of 50 tons per sq. in., a Brinell hardness as high as 400, and an electrical conductivity of 25 to 30%. Detailed information regarding this alloy is contained in an admirably produced booklet, which is available from Mallory Metallurgical Products, Ltd., 78, Hatton Garden, London, E.C. 1.

A new solder has been developed by Fry's Metal Foundries, Ltd., which has a lead base and contains only 5% of tin. Its use will effect considerable savings in the use of tin, since normally solder absorbs about 20% of the national consumption. This new solder has a melting point 50° C. higher, and its strength when used to lap-joint steel is 2·3 tons/sq. in., compared with 2·95 tons/sq. in. of a 40% tin solder. Full information is available on application to Tandem Works, Christchurch Road, London, S.W. 19.

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Control of Iron and Steel

THE principal purpose of the many orders issued by the Ministry of Supply, with respect to the control of iron and steel, is to conserve available supplies for use in the more important directions. Normally, the manufacture of carbon and alloy steels is adjusted by supply and demand, and although there are frequently considerable fluctuations in demand, the capacity of producers is rarely exceeded. Under present conditions, however, the demand for iron and steel far exceeds capacity, and the obvious solution to the problems arising from this cause is to organise distribution so that supplies will be available according to the urgency and importance of the demand.

INDEX TO VOL. 25.

In view of the paper shortage and to avoid waste, it is not proposed to issue an index with the May issue of "Metallurgia." Readers who desire an Index to Vol. 25 are asked to advise us, as copies will be prepared and despatched on request.

Although supplies of raw materials used in the manufacture of various carbon and alloy steels, and not available in this country, may be, and frequently are being, shipped in quantities comparable with peace-time imports, the demand is so large that supplies must be used to the best advantage, and although organised control can do much to ensure that these raw materials are conserved for vital purposes, both manufacturers and users of carbon and alloy steels can do much to obtain the best use of the materials. However well organised a system of control may be, it can only be successful with the active co-operation of producers and users; especially is this true of users.

The majority of readers will readily admit that, before this war, users of alloy steels were very extravagant in their choice, and frequently stipulated that the alloy steel should have properties far in excess of the service requirements of the part demanded. Even in peace time there is no justification for such extravagance, while in this period of national emergency it should be regarded as a sin. This question of special and alloy steels has been carefully considered by a committee of experts, embracing representatives of both manufacturers and users, and, as is described in greater detail elsewhere in this issue, the results achieved will be a very useful guide to users and greatly assist the effective application of alloying elements available in the most economical manner.

As one alloying element becomes difficult to obtain, the use of other elements is developed with a view to maintaining maximum efficiency in service from the materials produced. An instance of this is high-speed steel for which it has been customary to use a high percentage of tungsten. Difficulties have been encountered in the transport of this raw material from China, and, to some extent, it is being replaced by molybdenum. It is to this type of steel that a more recent order is concerned, the effect of which is that high-speed steel, whatever the tungsten content, can now only be acquired under the authority of a licence issued by the Iron and Steel Contro'. By the previous order, a licence was only necessary when the tungsten content was 14% or over. Existing licences for high-speed steel issued under the old order remain in force until May 1 next, on and after that date licences will automatically be valid only for high-speed steels containing up to 6·5% of tungsten.

The fact that goods made of raw materials in short supply owing to war conditions are advertised in "Metallurgia" should not be taken as an indication that they are necessarily available for export.

For the guidance of makers of small tools and other users of high-speed steel, in connection with the introduction and treatment of "substitute steels" to replace high-tungsten high-speed steels, an explanatory memorandum has been issued by the Ministry of Supply, compiled by agreement between the Iron and Steel Control (Director for Alloy and Special Steels), the Controller of Jigs, Tools and Gauges, and manufacturers of high-speed steel. "Substitute steels" are indicated by the following two typical grades:

	"Substitute 66." %		"Subs 94." %
Molybdenum	5·0 — 6·0	..	8·5 — 9·0
Tungsten	5·5 — 6·5	..	3·5 — 4·5
Chromium	4·0 — 5·0	..	3·5 — 4·5
Vanadium	1·25 — 1·50	..	1·25 — 1·50

These high-speed steels are not by any means new, as molybdenum high-speed steels have been developed and applied in practice for some years, but the technique applied in the working and treatment of these steels differs in some respects from that for high-tungsten high-speed steels: for this reason, Mr. Linley, who has had special experience with "Substitute 66," has prepared detailed information, which is published elsewhere in this issue. It will be noted that, according to their molybdenum content, these steels have a greater tendency to decarburise in forging and hardening than is the case with the standard high-tungsten high-speed steels; but with care this tendency should not cause trouble. Users who have little or no experience with these molybdenum steels will find Mr. Linley's article of special value. It is unfortunate that the term "substitute" has been applied, because experience in this country and in America has proved that the "66" type corresponds closely to standard 18% tungsten high-speed steel.

It should be borne in mind that the percentage of high-tungsten high-speed steel and high-speed steel containing cobalt permitted to be manufactured will be strictly limited, and will only be a small part of the total high-speed steel production; users are, therefore, advised in their own interests to familiarise themselves with the handling of molybdenum high-speed steels. and particularly with their heat-treatment.

Modifications in the choice of special and alloy steels will necessitate adjustments to normal practice. The more efficient use of raw materials forced upon manufacturers in the present emergency, however, will not only be advantageous to the national interests, but should prove invaluable to engineers and manufacturers when conditions are normal again.

Many more examples could be cited, the main point is, however, that an important factor in making the best possible use of alloying elements available concerns close co-operation between the metallurgist and the engineer; it is important, that the materials chosen should be governed by the particular requirements of each individual application.

Canada's Industrial Activity

THE acceleration in the tempo of Canada's war effort during 1941 had a profound effect on industrial activity which reached the highest level in history, according to a recent *Monthly Letter* of the Royal Bank of Canada. The rise in national income, the most comprehensive measure of productive activity within a country, was noteworthy, reaching an annual rate in November of $5,300 million, an increase of 10% with 1940.

An indication of the Dominion's war effort is to be found in the mounting total of government orders and contracts placed since the outbreak of hostilities. Directly or indirectly, the impact of these gigantic commitments which totalled more than $3.2 billion as at December 31, 1941, has been reflected in every branch of Canadian economic activity. By their very nature, however, the so-called heavy industries have enjoyed the greatest improvement. Production of iron and steel was at a record level in 1941. Capacity has been extended to meet war-time requirements, but current output is still short of demand. The situation is being met to some extent by imports from the United States, but with a serious shortage in that country drastic curtailment of civilian utilisation may prove to be the only solution.

In the main sections of industry capacity operations accompanied by new production records seem assured for the duration of the war. Manufacture of aircraft is now at a high weekly rate, with employment in the industry totalling 30,000 as compared with fewer than 1,000 before the outbreak of war. The Dominion's shipbuilding programme, providing for the construction of 100 freighters aggregating nearly a million tons, has been making favourable progress. During the year under review the chemical industry was engaged in diverting more and more of its facilities to the output of munitions and chemical products necessary to the prosecution of the war effort. During the year the production of high explosives and other war-time chemicals exceeded that of the four years 1914-18.

Mining continued to play a vital role in Canada's war effort, preliminary estimates placing the value of the 1941 output at an all-time high level of $553.9 million, despite relatively low prices for base metals. Gold production was slightly higher, while the output of silver showed a decline. Gold has retained its position as the largest single source of United States exchange, but it is being clearly recognised that further expansion in base metals production is an immediate necessity. Supplies of special equipment and structural steel are difficult to obtain, although material and labour costs have been stabilised by recent price-ceiling regulation.

Data relating to non-ferrous metal production are not available, due to war-time regulations. An analysis of exports indicates, however, that output was substantially higher in 1941 : total base metal shipments were valued at $244 million. Domestic consumption was also on a higher level. Contracts to supply the British Government with the major portion of Canadian base metal were recently renewed for a third year, with prices adjusted to the higher costs prevailing, and shortages in the United States have created a demand for a part of Canada's surplus production at somewhat enhanced prices. The nickel-copper industry is engaged in an expansion programme involving the addition of some 20% to productive capacity. Demand from Great Britain and the United States will undoubtedly continue at record levels ; except for scrap, these countries are almost wholly dependent on Canada for their supplies of nickel.

Intensive search for deposits of strategic metals, such as manganese, tungsten, chromium, tin, antimony, mercury and iron ore is at present being carried on with satisfactory results. Due to the paramount importance of aircraft in the present struggle, aluminium production has shown tremendous gains. To date, Great Britain has absorbed the larger portion of Canada's output, but, in the summer of last year, the United States contracted for a total of 340,000 metric tons to be delivered not later than 1944. This compares with a total output in 1940 of only 85,000 tons.

Stripping copper "tops" and nickel "bottoms" in the Orford mill, International Nickel Co., Ltd., at Copper Cliff, Ontario.

The High Efficiency Cyanide Copper Bath

THE high-efficiency cyanide copper process was introduced in 1938 by the du Pont Company, under the name of "High-speed Copper," to provide a method for the rapid application of smooth, bright, heavy deposits from cyanide baths. Before the introduction of this new process, the rochelle salt-bath was probably the most satisfactory cyanide copper bath for applying smooth, fine-grained copper deposits prior to bright nickel plating. Plain cyanide copper baths have also been used extensively for conventional electroplating purposes. The recognised deficiencies of the rochelle and other dilute type cyanide baths are :—

A. Variations in anode and cathode efficiencies (30 to 70%) make solution control difficult, promote cyanide decomposition and result in objectionable gassing.

B. Limited cathode current densities, together with low cathode current efficiencies necessitate unreasonably long plating times for production of heavy deposits in modern plating cycles.

These deficiencies are largely overcome by the high-efficiency cyanide copper process, which provides anode and cathode efficiencies of approximately 100%. However, in common with other bright plating processes, the high-efficiency cyanide process employs brightening and anti-pit agents and requires close adherence to established operating conditions with respect to bath composition and temperature.

This new bath, which is described by H. L. Benner and C. J. Wernlund,* comprises a high concentration of sodium cuprous cyanide, low "free" cyanide, a relatively high hydroxide content, a high operating temperature (80° C.), addition of brightening and anti-pit agents, and rapid electrode film replenishments by agitation of the work and the solution. Cathode current densities as high as 11 amp./dm.^{2} can be used.

A/B Svenska Metallverken, the Swedish manufacturers of finished and semi-finished non-ferrous metal goods, announces a somewhat decreased net profit of 1,180,000 kr. (£69,000) for 1941. The payment of an unchanged dividend of 8% is recommended.

* The Electrochemical Society, Oct. General Meeting, 1941.

Special and Alloy Steels

By Dr. W. H. Hatfield, F.R.S.

There is no justification in peace times for the extravagant use of raw materials, but at this time of emergency it should be regarded as a sin, and the best possible use should be made of the tonnages of alloying elements available for the manufacture of those steels that must possess properties superior to those of straight carbon steels in order to give the service required from them. The author discusses special and alloy steels in the light of present conditions, and refers to data presented in a simplified form designed to enable users to comprehend the possibilities of a range of types. Particularly, attention is directed to the economical use of these steels and to the careful salvage of their scrap.

THE writer welcomes the invitation to contribute an article dealing with the availability and application of special and alloy steels. Practice has long confirmed the extreme value of the influence of elements when added to the composition of steel. The addition of large percentages of selected elements produces the capacity of resistance to corrosion of various types, confers a resistance to scaling, and increases strength at elevated temperature, can induce a high coefficient of thermal expansion or can produce non-magnetic conditions, and other important changes in the chemical, physical and mechanical properties. High-tensile strengths with the maintenance of reasonable ductility, great wear resistance, and the facility for producing hardened surfaces, are all aspects of the properties which are effectively catered for. It is true too that, did space allow, a most interesting article could be devoted to the various types of these special steels, which have come into existence since the 1914–1918 war. Some of the special characteristics can only be obtained by the addition of large quantities of selected elements, and for such purposes, alternatives are not possible. This article can, however, be usefully devoted to those regions of the field where, when the incidence of the availability of the alloying elements varies, alternatives can be employed. For instance, it is not generally appreciated, but it is a fact nevertheless, that the initial application of alloys to steel metallurgy was for the purpose of improving armament materials : also such applications related to heavy masses, and therefore fairly rich alloy compositions were employed. It was a natural evolution that such materials should be applied to multitudinous purposes based on much smaller dimensions, and it is an interesting reflection that if the original development of alloy steels had been built up from the lighter end rather than from the heavy end, practice would have been based much earlier on steels of lower alloy content. There undoubtedly are to-day a vast number of applications of alloy steels where, owing to the dimensions, steels containing much lower alloy contents could be employed, and the same mechanical properties be obtained.

The main alloying elements with which we are concerned in order of importance are manganese, chromium, nickel, tungsten, molybdenum, cobalt and vanadium. Whereas hitherto the manufacturing concern producing alloy steels has been able to buy just as much of any element which it is desired to incorporate in a steel, the present extended emergency has produced a very different state of affairs and has necessitated reconsideration of the world's supplies and the availability of these different elements. War on the present gigantic scale has led to a demand for the alloying elements far in excess of their availability. It must always be remembered that availability has rested upon commercial enterprise and the demand at a given moment has resulted in the production of a given tonnage. Producers either as regards mining of the ore, or the reduction of the metal have necessarily for commercial stability had to restrict their efforts within reasonably narrow limits of the actual demand. It is practicable in one's own territory to develop more mines and more reduction units to make up deficiencies, provided, of course, that the necessary mineral deposits lie in such territories. Long periods of time are, however, necessary to bring into play such new facilities, and it is therefore essential to make the best use of the tonnages available and to apply the alloys as effectively as is possible according to the incidence of availability.

Defining Types

To the technologist whose principal hobby is the production of alloy steels of requisite properties the present emergency is a time of great opportunity. In this country the first step that was taken was to review the whole field of alloy steels consumption, and to schedule the various compositions and properties which apply to different uses. The outcome of the effort was very interesting in that it was found that, whilst in Britain the manufacture of special and alloy steels was governed by well over 2,000 specifications, the whole of the materials could be and were defined into 85 types or categories. The basis for the definition in each case was not necessarily the same, but it was either on the basis of composition, tensile strength, special properties or applications. Nevertheless, this classification, known as the T.A.C. classification, was of considerable value in focusing the whole position.

The technical difficulties militating against simplicity of exposition are obvious, and are for example well exemplified in the case of the hardened and tempered condition of alloy steel to give a tensile strength of 60–70 tons/sq. in., in which instance the same properties within a large range of sizes can be obtained either from a manganese steel, a chromium-molybdenum steel, a 1½% nickel-chromium-molybdenum steel, a 2½% nickel-chromium-molybdenum steel, a 3% nickel-chromium, 3% nickel-chromium-molybdenum, 3½% nickel-chromium-molybdenum steel, or a 3% chromium-molybdenum steel. Other steels also would produce the results.

It is true that the manganese-molybdenum steel would only give the properties up to dimensions somewhat in excess of 1 in. diameter, but the chromium-molybdenum steel might be used to 2½-in. diameter, whilst some of the latter mentioned steels would give the necessary properties up to a section of 6-in. diameter. It is, however, clear that in view of the extremely large tonnage of the smaller sizes it is not technically efficient to employ for these smaller sizes the material necessary for the heavy sizes. Another outstanding instance is the question of tool steel, and it is clear that the appropriate element to employ is the one in the greatest availability which will produce the desired result.

Without saying more at this stage of the consideration of the subject, the writer would indicate that as regards general engineering steels the work was crystallised into the form of the British Standards Specification 970/1941, known as the "En Series Wrought Alloy Steels." It was invaluable in the present emergency to have in being

the British Standards Institution under the direction of Mr. Le Maistre, which body formed an effective national focus and means of making known to the engineering world with suitable readiness the deductions with which we are dealing. In B.S. 970/1941 (which, by the way, is coming out almost immediately in a further increasingly useful form), will be found effective working specifications covering the whole requirements of wrought alloy steels for purposes up to 6-in. diameter, and in this document on pages 19 and 20 will be found what the writer considers to be a useful table indicating the possibilities for workshop practice of the various types of the steels most readily available.

This table shows that, over a wide range of sizes and for certain tensile strengths, carbon steels with appropriate manganese content function well; and it should be borne in mind that as regards small dimensions—i.e., up to 2-in. diameter—hardened and tempered carbon steels with the appropriate manganese content will consistently give mechanical properties which a material proportion of the users have probably deemed were only to be obtained from quite richly alloyed steels. When this most simple form of composition is exhausted in its possibilities, then one resorts to the introduction of the molybdenum content to what is approximately a $1\frac{1}{2}\%$ manganese steel, and when that is exhausted the next step is to go on to the chromium steels, then to the chromium-molybdenum steels. It is shown that only in the heavier sections is it essential to use nickel, which having in mind the present position as regards supply, is fortunate and cannot be too widely known.

British Standards 971

To facilitate a complete understanding and use of British Standard 970/1941, the Institution have published a further document, British Standard 971, which deals in a fairly complete manner with the steels corresponding to T.A.C. numbers 1 to 33. In this document many matters are usefully discussed, as, for instance, the very much debated question as to what is a reasonable range of composition to which a steelmaker should supply—it being pointed out that whilst when the supplier does the hardening and tempering a fairly wide range of composition is permissible, yet when the consumer does his own heat-treatment there are practical limitations in the degree of uniformity which can reasonably be expected. For instance, experimental error in chemical analysis, heterogeneity of a normal character in the material, and differences in process, all have to be taken into account. These matters are very important when the production of such materials is being undertaken by an increasing number of manufacturers coming new to the field. There is a companion aspect of matters which should also be mentioned—namely, that not only do the aforementioned variables cause a fairly wide range of compositions to be inserted in specifications, but there is the additional fact that both the user and the producer of the steel will quite rightly have his own ideas as to exactly what composition is essential to achieve the mechanical properties which are sought. This document also contains an interesting statement on mass effect and ruling section which should be read and studied most carefully by all users of alloy steels. After a section presenting the T.A.C. Schedule up to No. 33, it will be found that many pages are devoted to conveying to the engineering world, practical experience of steel-makers as regards obtaining the various mechanical properties from these different steels, and a study of this section results in a detailed implementing of the possibilities of the steels outlined in the specifications in B.S. 970/1941. Lastly, the document contains an effort to cross-reference a multitude of specifications to the simple schedule which has now been created.

The two British Standards documents to which reference has now been made cover a field of technology in detail which only those who have really spent many years in studying can quickly assimilate. Yet it is essential and possible in the present simplified form, for engineering users readily to gain a complete comprehension of the possibilities of these materials.

Material v. Service Requirements

It is worth while enunciating that it is the ultimate mechanical properties of the steel of which the part is made which matter, and not the composition. Mechanical testing is now so complete that checks can so readily be made for the satisfaction of anybody interested in order to confirm that the mechanical properties necessary for a given part are produced satisfactorily in the alternative and probably low-alloy steels. The trial and error method of practice has, however, provided a rich background of knowledge which readily facilitates the successful development of the present efforts both in Great Britain and America, only to use those elements and in amounts which are essential for obtaining in practice the given results. One matter might here be mentioned to which the writer has frequently had reason to refer, and that is the disabilities resulting from the user calling for material of higher properties than he really requires: for instance, if he calls for a 65 tons tensile, where a 50 tons tensile strength would in practice meet his requirements; where he calls for a high-notched bar value where a lower-notched value would not matter, and also in such cases as where steel is operating at a temperature and he may inform the producer that the temperature is 100° higher or more than in reality is the case; when a little more meticulous appreciation of the fact that the higher demands forces the producer on to a more liberal use of alloys than is really necessary. Another point that the writer cannot emphasise too strongly is that in his long experience of investigating the failure of parts in mechanisms covering a very wide range of type, well over 90% of the failures have not been due to defective or unsatisfactory material, but have been due to faults in design—i.e., due to an inadequate balancing in the designer's mind of the capabilities of the material and the service which he was requiring of it; often the metallurgist is called upon by modifications in the properties of the material to remove disabilities of design, such as sharp corners producing a concentration of stress.

If, in this period of abnormal demands for alloy steel of one type or another, success is to be achieved as regards economical application, it is essential that the whole consideration of the problem shall begin at the designer's end. The steelworks producing the alloy steel is not generally aware of its specific application and therefore it must remain the responsibility of the user to employ these materials to the best advantage.

The Salvage of Alloy Steel Scrap

The writer feels that he cannot end an article of this character without referring to the need for the effective utilisation of the alloy content remaining in the scrap of various types produced during the fabrication from the ingot to the finished article. In a well-organised steelworks the initial scrap should be, and undoubtedly is, generally utilised internally. This is not always the case with regard to turnings and like scrap. It is not always appreciated what tremendous tonnages are involved in this particular aspect of the subject. The scrap produced in fabricating the ingot into a bar or a forging is understood; but it is not always understood that frequently even in producing a rough-turned product, the turnings are equivalent to quite a high percentage of the weight of the rough-turned forging. When, however, we arrive at the engineering works where the final machining is done, it is not an uncommon occurrence as regards articles produced at the present time for the actual turnings to weigh far more than the finished product. Every engineering establishment should ask itself whether it is playing its

part in suitably segregating according to composition what represents in the aggregate a very great tonnage of material. The weight of nickel, chromium, molybdenum and other elements which can be lost if some adequate system of salvage is not in operation, is quite large.

The writer feels that the present extended emergency will have a very beneficial influence to the advantage of all engineering establishments in that they will be, and should be, forced into a more complete understanding of the properties of the available special and alloy steels. The effects of present experience will undoubtedly have a very lasting result and post-war manufactures will have the advantage.

Physical Properties of Brass Cartridge Cases

The relatively low modulus of elasticity of brass combined with its fairly high range of elastic action is the reason for its successful use in the manufacture of cartridge cases, and this article discusses recent changes in their manufacture, resulting from the use of copper and zinc of greater purity. Brief reference is made to cold-working operations, and particular attention is directed to their effect on the physical properties of various parts of the case. Results from rolling and annealing tests on typical material are given, and reference is made to the finish of cases to prevent danger from corrosion.

IN normal peace times the manufacture of ammunition is given very little thought by the average person, though it is an essential requisite of defence requirements; in times of war, however, a new interest is taken in its manufacture by the people of the country at war; their chief concern being that the scale of manufacture be graded upwards to meet any possible requirements of the fighting forces. The term ammunition, as applied in a military sense, includes projectiles used with ordnance, rifles and similar weapons, together with the requisite propellants, fuses, cartridges, and means of ignition.

In the smaller calibres of ammunition the propelling charge is contained in a brass case known as a cartridge case which has a primer in the base to ignite the charge and is crimped to the base of the projectile. This type of ammunition is known as fixed ammunition, and is usual in calibres from 3·7 mm. to 4·7 in. The purpose of the cartridge case is not only to provide a means for igniting the explosive charge, but also to prevent the escape of gas generated from the charge. In the centre of the case head an opening is made to hold a primer, which is ignited by the blow of the firing pin in the gun mechanism. The metal around the primer hole must be sufficiently hard and elastic enough to ensure that there is no loosening of the primer and no gas leakage at the primer hole due to the pressure developed in firing.

The success of the breech-loading gun was dependent on the use of the cartridge case, but early attempts in the design and manufacture of these cases were unsuccessful until in 1870 General Berdan, of the U.S. Army, invented the centre-fire priming device, using a brass shell as the container. According to Pratt,* the brass case invented at that time is almost identical in principle with the present brass cartridge cases. The difficulties encountered in the initial stages centred round the material used. The wall of the case must be elastic enough to expand under pressure, and make a tight seal against the wall of the gun. The power developed by the propellant charge is sufficient to burst any case unless it is properly supported by the breech wall: thus, the clearance between the case and the breech wall must be carefully controlled so that a great portion of the strain is taken up by the steel of the breech wall. The relatively low modulus of elasticity of brass, combined with its fairly high range of elastic action, is the reason for its successful use in a cartridge case. The low modulus permits the quick expansion of the case to the breech wall, and subsequent deformation of the case is controlled by the deformation of the steel. Its elastic limit is sufficiently high to permit this deformation to take place without permanent set, and therefore the case returns to its normal size when the pressure drops. These requirements are necessary to produce the gas seal and the easy removal of of the case after firing.

Another factor which seems to enter into the problem is the relatively high thermal conductivity of cartridge brass. A great deal of heat is developed by the combustion of the charge, and it is probable that the dissipation of this heat through the brass case wall is sufficient to reduce the net expansion following the firing. A case of low thermal conductivity might attain such temperatures as to prevent its rapid removal from the breech, and might even result in softening of the case wall. It will be apparent that the substitution of other case materials for brass is dependent on several factors which are difficult to control so as to meet requirements. Such substitution, therefore, involves careful study and experimentation.

Changes in Manufacturing Process

The original centre-fire cartridge case was made of yellow brass containing about two parts copper and one part zinc. The present cartridge brass is of the same nominal composition—viz., 70% copper and 30% zinc, in some instances a small percentage of tin is added to improve the mechanical properties of the alloy, but the purity of copper and zinc are so much greater in modern practice that the mechanical properties of the alloy and its ability to withstand deformation are much improved. Until as recently as five or six years ago, all cartridge brass had been processed entirely by cold-working operations. The zinc obtainable up to that time contained appreciable quantities of lead, cadmium, and iron as impurities. When zinc of high purity was made available it was found that cartridge brass made from such zinc was capable of hot working without difficulty. Subsequent development of heavy hot-rolling mill and casting equipment suitable for the production of large-size bars made hot-rolling the natural procedure for the initial breaking down operations. The great purity of the brass which has made possible these changes in mill operations has also made possible a decreased number of cold-drawing operations in the production of the case.

Modern practice in the production of cartridge brass starts with the casting of bars 500 lb. to 1,000 lb. in weight which are broken down by hot rolling. The method by which the final size is reached will depend on the particular case which is to be made. For some cases the rolled bar will be machined to size, for others a cold-rolling operation

* From a contribution by R. S. Pratt to the Metals Engineering Division and presented at the annual meeting of the American Society of Mechanical Engineers, 1941, and published in *Mech. Eng.*, **64**, No. 2, 1942, pp. 119–122.

will be made, followed by annealing to produce the specified grain size for cupping. Blank discs of the proper diameter are then cut from the rolled bar.

The discs or circles are cupped by forcing them through a die of smaller diameter and constructing the circumference of the blank around the punch. In the last Great War the case for the 75 mm. field gun was made by cupping a circular blank and a series of five or six redrawing operations. The same case is made to-day in only three draws—a considerable saving in time and consequent increase in production. This improvement is due to the improved purity of the alloy used as well as to improved presses of greater power. The same factors have made it possible to produce cases for some of the larger and more powerful guns, which were not economically possible a decade ago.

In addition to the reduction in the number of redrawing operations, improvements have also been made in the finishing operations, such as heading the base, and the local mouth-annealing and relief-annealing operations. The improvement in heading has been due largely to increased press capacity and improved die steel, so that full advantage may be taken of the plasticity of the brass. The machining operations have been speeded up by more modern machine tools, cutting tools, and lubricants. In order to take advantage of the improved equipment improved annealing equipment has been provided in which the properties of the annealed brass may be accurately controlled within fine limits. Taper and mouth annealing are almost entirely done by automatically controlled salt-bath anneals. Relief annealing, which is an extremely sensitive anneal at temperatures in which heat transfer is rather slow, is now done in forced-convection furnaces or salt baths in which the time of exposure can be appreciably reduced and the results more accurately controlled. Not the least of the steps taken to increase production has been the application of modern production methods for the inspection of the finished cases. All the dimensions of every cartridge case must be controlled within very narrow tolerances, and the checking of each dimension is a lengthy process.

A typical cartridge case is sketched in Fig. 1. The physical properties and temper of the various parts will vary slightly, depending upon the gun and the explosive charge the case must contain. The tensile strength and hardness data given in this illustration relate to the requirements for the 75 mm. M18 fixed ammunition case. From the physical properties specified it is apparent that the various portions of the case should be processed differently in order to meet the requirements. Test samples from positions indicated gave the following results: At A a tensile strength of 52,000 lb./sq. in. with elongation 51·6% in 2 in.—the structure comprises a fairly fine grain as a result of the mouth anneal. At B the tensile strength was 53,950 lb./sq. in., and elongation 48·5% in 2 in.—this sample possessed a very fine-grained structure, resulting from the taper-annealing operation; a slight amount of cold work may be left from the tapering operation. The points C and D are in the severely cold-worked condition, resulting from the final drawing operation, and this has a considerable effect on the properties of alloy in these positions, the tests for C giving a tensile strength of 96,800 lb./sq. in. with an elongation of 7% in 2 in., and for D 101,900 lb./sq. in. and 5% in 2 in. The point E in the base is also in the cold-worked condition as a result of the heading operation, but it is less severely worked than at points C and D, and tests gave a tensile strength of 62,100 lb./sq. in. and an elongation of 30% in 1 in.

Structure and Properties

In order to withstand the bursting effect of the explosive charge, cases are made with a heavy base and with a side wall, which is heaviest at the base, tapering in thickness as well as diameter towards the mouth. The tapering of the wall thickness would normally result in less cold-working of the wall at the base. Actually, the opposite is required, and the tapering of the wall from the inside is carried through the entire series of draws. Therefore, the final draw is such as to produce a cylindrical shell with a heavy rounded bottom and a side wall tapering gradually in thickness towards the mouth. The heavy bottom has received but little cold work in the last draw, but the side wall is severely cold-worked in the heaviest portion near the base. In order that the bottom will have the necessary strength, it must be cold-worked to approximately its final shape. After heading the case cannot be annealed in entirety, and such other anneals as are necessary must be local anneals to prevent softening of the base and lower side wall.

While the case shell, as drawn, tapers in the side wall, it is not possible to produce a diametrical taper by such an operation. It is therefore necessary to introduce a tapering operation to produce the general taper and the reduced mouth diameter characteristic of most cases. In order to perform the tapering and reducing operation it is necessary in most types of cases to anneal the open end. This is done by immersing the open end and about half the length of the case in a molten salt bath. The case is then reduced to its final form and, if the case is to be used for fixed ammunition, it is again annealed.

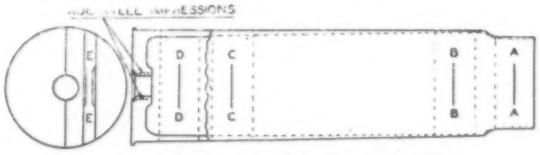


Fig. 1.—Sketch of typical cartridge case.

TYPICAL SPECIFICATIONS FOR CARTRIDGE CASES, 75mm.—M18.
PHYSICAL PROPERTIES.

Position.	A.	B.	C.	D.	E.
Minimum tensile strength, lb./sq. in.	45,000	45,000	65,000	65,000	60,000

Primer Hole Hardness.
Rockwell—B65 ($\frac{1}{8}$ in. ball 100 kg. load).

Some consideration of the effect of these various cold-working and annealing operations is indicated in the accompanying graphs. Figs. 2 and 3 are characteristic curves which illustrate the properties of cartridge brass as they are affected by these processes. The curves in Fig. 1 are produced from experimental data obtained by cold-rolling a strip of cartridge brass of various amounts and selecting samples for test after each pass through the rolls.

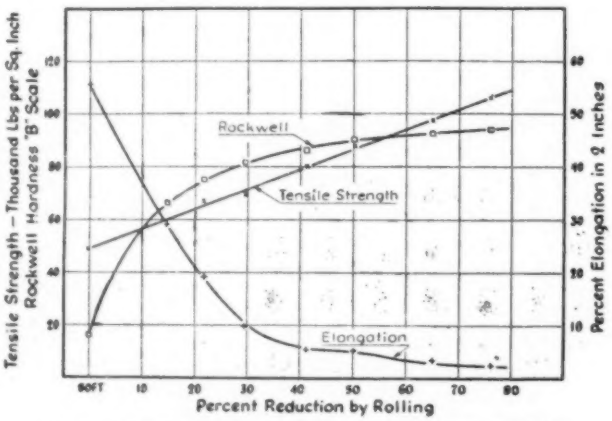


Fig. 2.—Rolling characteristic curves, sheet cartridge brass annealed at 0.100 in. gauge—grain size 0.058 mm.

The samples selected were tested for tensile strength, hardness and elongation per cent., and the results plotted against the percentage in reduction in thickness by cold-rolling. The curves show clearly the increase in tensile strength and hardness which results from cold-rolling. The sharp decrease in ductility shown by the drop in the

elongation values between 0 and about 20% reduction indicates the reason for the annealing which is carried out between the various drawing operations and before the tapering operation.

Fig. 3 shows the effect of annealing. These curves were developed by selecting a sample of cold-worked material which had been uniformly processed up to this point. The specimen was cut into several short lengths, which were annealed individually by exposing to various temperatures for 1 hour. The samples were air cooled and then tested for tensile strength, hardness, elongation, and grain size. The results of these tests plotted against the annealing temperatures produce the data and curves shown.

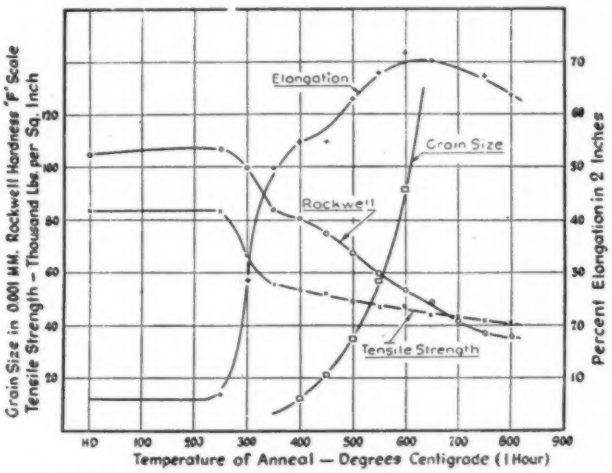


Fig. 3.—Annealing characteristic curves, sheet cartridge brass rolled 50% hard to 0.050 in. and annealed at indicated temperatures.

It will be noted that temperatures below about 250° C. do not produce any change in properties except perhaps a slight increase in hardness. At about 300° C. the brass softens rapidly and recrystallisation occurs. These changes are indicated by the rapid increase in elongation or ductility and the decrease in hardness and tensile strength. Further increases in annealing temperature produce a somewhat further decrease in hardness and tensile strength and a very considerable increase in elongation. With the particular cold-worked material used in these tests, recrystallisation is about complete at 350° C. and the alloy structure consists of a fine equiaxed grain, solid-solution structure. As the temperature is increased the size of the grains increases rapidly, and this increase in grain size accounts for the increase in elongation. The point at which recrystallisation occurs, the initial recrystallised grain-size and other details of the annealed properties are affected by the amount of cold work which the brass has received.

Relieving Internal Stresses

The mouth of the case is the only portion that is entirely free from internal stress as produced, because of this it is necessary that the entire case be relief-annealed. Experience with season cracking of cartridge cases in the last Great War clearly demonstrated the danger from internal stresses. The advantages of relief-annealing were also demonstrated at that time. Such annealing is rather critical, however, in that the use of high temperature is likely to result in softening of the cold-worked portion. The most efficient relief-annealing temperature is one just below the temperature at which softening begins; it is apparent from the data in Fig. 3, however, that this temperature will not be the same for all parts of the case, and, in order to be safe from softening, a temperature is selected which will relief-anneal the most severely cold-worked portion without softening. The mercury-test results then indicate whether the temperature-time cycle selected has been sufficient to reduce the internal stress below the danger point. The results of tensile tests or hardness tests together with micro-examination will indicate whether softening has taken place.

At relief-annealing temperatures 250°–275° C. heat transfer by radiation is slow, and there is a likelihood of overheating parts of the charge in an ordinary muffle furnace. More recent practice has been to use convection heating methods, which assure rapid heating and reduce the possibility of overheating. Salt baths are also very satisfactory for relief-annealing, except for the difficulties in removing salt, which may solidify on the case. Care in washing will remove such salt.

Finish

In addition to the requirements involving dimensions and physical properties, there are those requirements which seek to ensure the satisfactory storage of the cases either loaded or unloaded for indefinite lengths of time. Considerable care is given to cleaning the case, to ensure its freedom from all acids or alkalies likely to cause corrosion of the metal or attack the charge within the case. It is unusual in works accustomed to producing finished brass articles to realise that cases are regularly produced with varying degrees of surface oxidation. There is no objection to such oxidation comparable with the objections which would be raised if attempts are made to remove the oxidation. The acids or alkalies necessary to remove the oxide are a much greater source of potential danger. Specifications specially prohibit the use of sodium bichromate as a means of brightening the surface. Inferentially, such pickling solutions as sulphuric acid or nitric acid are also prohibited, as well as cyanide solutions.

The finished case must be chemically clean from acids or bases, oil, grease or dirt, although it may be tarnished with varying degrees of oxidation. This degree of surface cleanliness can only be attained by very careful rinsing operations. Mild caustics are necessary to remove the cutting lubricants used in machining the base, and the rinsing is necessary to remove the caustic. The final rinse is usually of hot water, so that rapid evaporation takes place which prevents water stain. It is desirable that the rinse waters be soft, if possible, so that no salt residue remains after evaporation in the final rinse.

Sir William Bragg, O.M

JUST after going to press with the last issue of this journal, we were informed of the death of one of Britain's greatest scientists—Sir William Bragg. Born near Wigton in Cumberland, where he attended the village school, his outstanding brilliance enabled him to pass through various stages to Cambridge University, where he took his degree in science, specialising in mathematics. His first appointment was to the Chair of Physics and Mathematics at Adelaide University, where he remained for 22 years, and it was only after this period that he began to interest himself experimentally in crystal analysis and X-ray spectrography; in this field are his greatest contributions to science. He was a great advocate of the value of scientific research to industry, and he was honoured for his services in the last great war, when he was made a C.B.E. in 1917 and advanced to K.B.E. in 1920. In 1931 he received the O.M.

Although Sir William held numerous positions, including President of the Royal Society from 1935 to 1940 and Director of the Royal Institution, he remained a charming and lovable man, and had the rare ability of making the most abstruse subject interesting; his lectures to children in which he demonstrated scientific principles were not only informative, but invariably thrilled his audiences. Apart from the honours conferred upon him which have been mentioned, he was the recipient of many other awards, notably the Barnard Gold Medal and the Romford Gold Medal of the Royal Society; the Copley Medal and the Gold Medal of the Franklin Institute of Philadelphia; the Faraday Medal of the Institution of Electrical Engineers; and the Platinum Medal of the Institute of Metals.

Magnesium and its Alloys

By F. A. Fox, M.Sc.

Though first isolated as a metallic element early in the 19th century, magnesium was not used structurally until 1909. Like most pure metals magnesium is soft and has low mechanical strength, and only after continuous research with alloying elements to improve its strength characteristics was it possible to take advantage of the very low specific gravity of this metal as a structural material. As a result of this work a range of magnesium-base alloys has been developed which are in regular production for many applications. Magnesium-base alloys have characteristics which make them distinctive among metals. They are the lightest structural alloys yet known, being two-thirds the weight of aluminium and one-fourth that of iron or steel. They also possess excellent machinability, toughness and fatigue endurance. They have a high strength/weight ratio, possessing a strength, on an equal weight basis, equal or superior to most other metals. Alloys of this type in commercial use were discussed by Mr. Fox at a recent meeting of the Manchester Metallurgical Society, to whom we are indebted for permission to present his views, in a slightly abridged form, in this article.

A GREAT number of magnesium alloys have been patented since 1909; the majority, however, have been found to be valueless for practical purposes, and alloys which have been commercially selected have numbered about a dozen. A list of commercial casting alloys used during the past fifteen years, together with American equivalents, is given in Table I. The data given represents average values, and the difference between the British and American values may be due to difference of opinion on what constitutes average results. In dealing with magnesium alloys, as indeed with other materials, it is always desirable to regard the mechanical properties statistically, and to think of the most frequently obtained properties for a given alloy. Conclusions must not be based on only a few mechanical tests.

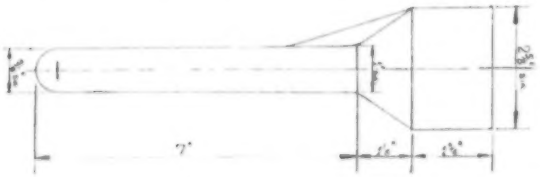


Fig. 1.—D.T.D. standard test-bar casting.

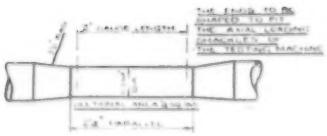


Fig. 1a.—D.T.D. machined test bar.

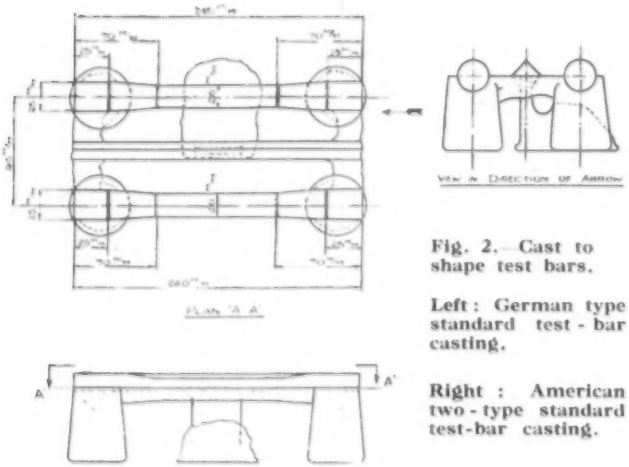


Fig. 2.—Cast to shape test bars.

Left: German type standard test-bar casting.

Right: American two-type standard test-bar casting.

In comparing American and German test results with those obtained in this country, it must be remembered that British practice is to use D.T.D. type test-bar casting, shown in Fig. 1; this is machined to the test-bar form shown in Fig. 1A. Foreign practice is to use test bars which are cast to shape: typical American and German test bars are shown in Fig. 2. The difference in mechanical properties between these two types is not great, but tests have shown that the D.T.D. type test bar gives lower mechanical property figures than the cast-to-shape bar.

It will be seen from Table I that the alloys which are in current use are, with the exception of the alloy covered by D.T.D. 140A, based on the magnesium-aluminium-zinc system. The two casting alloys most widely used in this country have compositions corresponding to that of Elektron A.8 and Elektron AZ.91, which approximate to the 8% and 10% aluminium alloys, respectively. Some of the factors which have led to the selection of these two casting alloys, as the most useful magnesium-base materials for engineering applications, are shown in Fig. 3, which indicates the change of mechanical properties of sand-cast magnesium-aluminium alloys as a function of the aluminium content. A similar diagram for the alloys as solution treated is shown in Fig. 4.

These graphs are of considerable interest: they show that the 8% aluminium alloy has a useful combination of properties in both the as-cast and solution-treated condi-

tions. This alloy has a maximum elongation as solution-treated, and has satisfactory proof stress as cast. The 10% aluminium alloy has a slightly higher proof stress as cast, although the ultimate tensile strength is somewhat lower than that for the 8% aluminium alloy: in the solution-treated, however, the ultimate tensile strength is a maximum for the whole system, and the percentage elongation is still good.

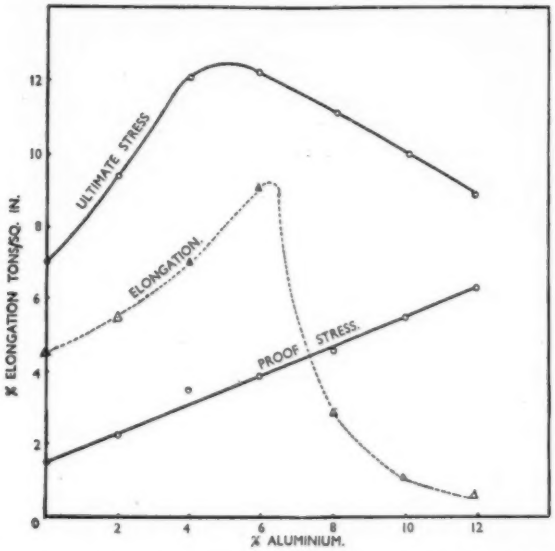


Fig. 3.—Mechanical properties of aluminium containing "as-cast" magnesium alloys as a function of the aluminium content.

It will be noted in the table, however, that the Elektron alloys A.8 and AZ.91 are not binary alloys; they contain small amounts of zinc and manganese. The effect of zinc on the mechanical properties is small, but the presence of a small amount of zinc possibly has a beneficial effect on the corrosion resistance, and it is thought to have a beneficial effect on casting characteristics. However, the presence of about 1% of zinc is believed to have a harmful effect on casting characteristics, particularly with those alloys of lower aluminium content. It has been found that high zinc contents tend to produce hot shortness, and special precautions have to be taken to ensure that thermal stresses are at a minimum in castings carrying a high zinc content.

The presence of about 0·25% manganese has been found to be highly beneficial from the point of view of corrosion resistance. When aluminium is the only alloying element, the effect of increasing amounts on the corrosion resistance is shown by Fig. 5. By analogy with its behaviour in other alloy systems, it might be expected that increasing the aluminium content would improve the corrosion resistance of the alloys. Actually, however, the corrosion resistance of the magnesium-aluminium alloys is worse than that of magnesium, probably because the aluminium phase produces local cell action, while the addition does not increase surface protection.

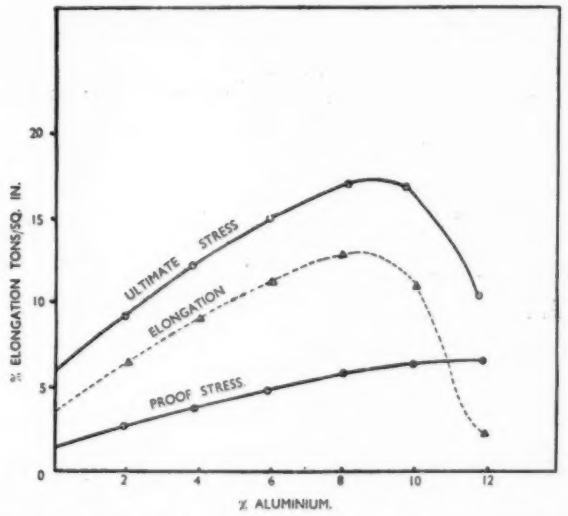


Fig. 4.—Mechanical properties of aluminium containing solution treated magnesium alloys as a function of the aluminium content.

When manganese is added to the aluminium containing alloys, in amounts of the order of 0·3%, conditions are changed; the corrosion resistance increases with increasing aluminium content, as indicated in Fig. 6. It is thought that the manganese contributes to the formation of an oxide layer, which improves the corrosion resistance. This

TABLE I.

COMMERCIAL MAGNESIUM CASTING ALLOYS USED DURING THE PAST 15 YEARS.

Alloy.		D.T.D. Specification.	Condition.	Average Tensile Properties.			American Equivalent.		A.S.T.M. Specification.	Average Tensile Properties.		
Elektron Designation.	Nominal Composition.			0·1% Proof Stress, Tons/sq. in.	Ultimate Stress, Tons/sq. in.	% Elongation on 2 in.	Trade Designations.	Nominal Composition.		0·2% Proof Stress, Tons/sq. in.	Ultimate Stress, Tons/sq. in.	% Elongation on 2 in.
CM Si*	Mg + 1·0 Si	—	Sandcast	2·3	6·3	3·0	—	—	—	—	—	—
AM.503	Mg + 1·5 Mn	140A	Sandcast	1·9	7·0	7·0	AM.403C Dow M	Mg + 1·2 Mn (max.)	B80—41T Alloy 11	2·0	6·2	5·0
AZ.31*	Mg + 3 Al + 1 Zn	59A	Sandcast	3·0	10·5	8·0	—	—	—	—	—	—
AZF*	Mg + 4 Al + 3 Zn	59A	Sandcast	4·5	11·0	7·0	—	—	—	—	—	—
AZG	Mg + 6 Al + 3 Zn	59A	Sandcast	5·0	12·5	5·0	AM.265C Dow H	Mg + 5·3/6·7 Al + 2·5/3·5 Zn	B80—41T Alloy 4	5·3	12·0	6·0
A.8	Mg + 8 Al + 0·4 Zn	59A	Sandcast	5·3	10·5	3·0	AM.241C Dow A	Mg + 7·0/9·0 Al	—	7·6	13·4	2·5
AZ.91	Mg + 9·5 Al + 0·4 Zn	136A	Sandcast	6·0	9·6	1·5	—	—	—	—	—	—
A.8	Mg + 8 Al + 0·4 Zn	59A	Diecast	5·8	14·0	8·0	—	—	—	—	—	—
AZ.91	Mg + 9·5 Al + 0·4 Zn	136A	Diecast	6·0	12·5	3·0	AM.230C	Mg + 9·0/11·0 Al	B94—41T Alloy 12	9·8	13·8	1·0

* These alloys are now seldom used. All the Al-containing alloys carry 0·15 — 0·40% Mn.

is most certainly the case in the alloy Elektron AM.503—used more as a wrought alloy than a casting alloy,—which is a simple magnesium-manganese binary alloy.

In the aluminium-containing alloys, it has recently been found that there is a safe lower limit for manganese, below which any manganese present confers only slight corrosion resistance. The lower limit has been found to be about 0·15% for alloys containing 5% to 10% aluminium, and it applies both in the protected and unprotected states. Figures are given in Table II showing the weight losses by corrosion for alloys containing 5% aluminium with differing amounts of manganese, both in the chromated and untreated conditions. It will be noted that the effect of insufficient manganese "shows through" the protection given by the chromate surface treatment.

TABLE II.

THE RELATIVE CORROSION RATES OF MAGNESIUM ALLOYS CONTAINING 5% ALUMINIUM AND WITH VARYING MANGANESE CONTENTS.

Nature of Test: 100 hours' immersion in 3% sodium-chlorine solution.

Specimens.	Relative Corrosion Rate of Alloys containing: Percentage Manganese.					
	0·015	0·10	0·19	0·33	0·38	0·50
Unprotected	100 Arbitrary Standard	3·2	1·3	1·1	1·1	1·4
B.A.E. Chromated	100 Arbitrary Standard	14	6·8	5·0	2·6	3·2

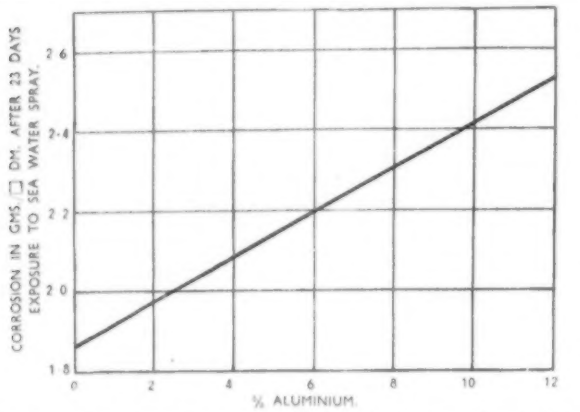


Fig. 5.—The corrosion rate of magnesium-base aluminium alloy as a function of aluminium content.

In addition to mechanical property and corrosion considerations, magnesium alloys, to be satisfactory in the foundry, must contain a reasonably high proportion of alloying element. The total linear contraction of magnesium is nearly 2%. Alloys containing over about 97% manganese are prone to severe local shrinkage, while the more highly alloyed materials have a lower total shrinkage and a wide solidification range, and can be readily fed and have better casting characteristics. A limit must, of course, be imposed on the addition of alloying element from other points of view, and at 12% aluminium the mechanical properties are beginning to deteriorate rapidly, the alloys becoming particularly brittle. The castability of the aluminium-containing alloys increases almost linearly with aluminium content, and for this reason the Elektron alloy AZ.91 is preferred for complicated castings to the Elektron alloy A.8.

Soundness

In contrast with aluminium alloys, magnesium alloys present no problem to the foundrymen as far as dissolved gases are concerned. Since the magnesium-oxide layer is readily permeable, gases which may be dissolved in the melt at high temperatures are easily expelled during cooling of the molten charge. However, since the aluminium and zinc-containing alloys have a wide solidification range there is a tendency for local porosity to be produced when a casting cools. This is produced naturally by the falling away of eutectic liquid retained within the "skeleton" of already solidified metal. This porosity occurs at those parts of a casting which are last to solidify, and such microporosity may therefore appear in bosses or thick sections of a casting. It is often readily possible by proper foundry technique to cause the microporosity to appear only in those parts of castings which cut off on fettling.

Certain alloys are more prone to microporosity than others; this tendency is consistently shown by the alloy containing 6% aluminium and 3% zinc, and it is now seldom used in this country, in spite of the fact that its mechanical properties are good. This alloy is still in wide use in America, where its mechanical properties and its slightly improved corrosion resistance—conferred by the presence of the 3% zinc—are considered attractive.

Influence of Impurities

The impurities which are usually made the subject of upper limit figures in British official and unofficial specifications are tin, lead, copper, iron, silicon. Of these, iron and copper are the most harmful as far as corrosion resistance is concerned, and it is also known that nickel in small amounts leads to rapid corrosion. It has been shown that the corrosion rate of highly purified magnesium in the presence of 3% sodium-chloride solution is very slow, and iron in amounts up to 0·017% has no effect on the corrosion rate; in quantities over this amount, however, there is a marked increase in the corrosion rate, and also that there is a definite upper safe limit for nickel at less than 0·0005%, and copper shows such a limit at about 0·1%. As far as binary magnesium alloys are concerned, it is noteworthy that investigations have shown where aluminium is present

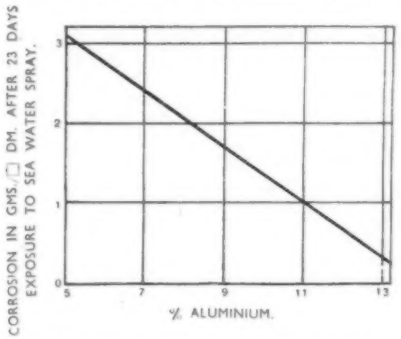


Fig. 6.—The corrosion of magnesium-base aluminium alloys (containing 0.5% manganese) as a function of aluminium content.

the safe limit for iron is greatly reduced, and at 10% aluminium the limit was found to be too low to be determined, even though analysis could be carried down to about 0·0002% of iron. When manganese was added to this binary alloy, however, the upper safe limit for iron was increased to 0·002%.

Although the elements, tin, lead and silicon, are sometimes held to a maximum for specification purposes, their influence, even in relatively large amounts (such as of the order of 0·4%) is not particularly harmful in the aluminium-containing alloys. It is, however, desirable to keep even these impurities down in order to ensure standard behaviour from the alloys under all manufacturing operations.

The binary alloy containing about 1·5% manganese has the best corrosion resistance of magnesium base alloys, but this property is at its best when the iron content of the alloy also is under 0·02%. The presence of aluminium is also detrimental, and contamination from aluminium scrap will lead to hot cracking of castings in this alloy, and will also cause them to crack if, during fabrication, they have to be welded into a larger structure.

Wrought Alloys

The wrought alloys which have been put forward during the past fifteen years are given in Table III, with American equivalents. The mechanical property figures given represent average values. In this country, the magnesium

alloys generally used in the wrought state are: an alloy containing 6% aluminium, 1% zinc, and 0·2% manganese for extrusion; one containing about 8% aluminium, 0·4% zinc, and 0·2% manganese for forging; while an alloy used for rolling consists of the binary magnesium-manganese alloy, containing about 1·5% manganese.

Sheet

An interesting development of the past two years has been the rolling of manganese containing sheet alloy, in this country, direct from cast slabs. The practice formerly adopted consisted in extruding a billet about 12 in. diameter into the form of a "blank" about 1½ in. thick. This was then hot-rolled, with various reheatings, to a thickness near the finished gauge, the last few thousands of an inch reduction in thickness being made cold. By producing a slab under carefully controlled casting conditions, the manganese alloy can now be rolled direct, and the mechanical properties of the finished sheet are as good as those of the same alloy which has been extruded as well as rolled.

An interesting point about the magnesium-manganese and other magnesium-base alloy sheet is that when directionality is present, all the properties are higher in a direction at right angles to the principal rolling direction than in the longitudinal direction. Where the rolling reductions are correctly applied, and reheats properly spaced, the actual amount of this directionality is small. On the other hand, in improperly produced sheet the directionality can be large; the following is an example of the difference between the two directions:—

Mg-Mn Alloy Sheet.	0·1% Proof Stress, Tons/sq. in.	Ultimate Tensional Stress, Tons/sq. in.	% Elongation on 2 in.
Longitudinal	6	13	5
Transverse	9	16	10

This magnesium-manganese sheet welds very satisfactorily with the oxy-acetylene torch, and ultimate tensile strengths of well over the A.I.D. requirements of 8 tons/sq. in can be obtained across an unhammered weld. It is essential, however, to use a suitable flux for welding alloy sheet and tube; this usually contains some constituent which acts as a solvent for magnesium oxide. It is desirable to avoid the use of a flux containing sodium salts, since these tend to cause obscuration of the work by their intense colouration of the flame.

The manganese-containing sheet will successfully withstand simple forming operations at room temperature, while severe forming or drawing is applied at a temperature about 350° C. At this temperature directionality effects are not important, and the sheet is readily workable, having great ductility in a wide plastic range. The temperature for working the sheet is not critical so long as it is reasonably well above the temperature of about 225° C., and as long as it is not so excessive as to cause rapid grain growth. The temperature of 225° C. is that at which in pure magnesium a new set of slip planes become operative and greater ductility is conferred on the material. This temperature is probably approximately correct also for the magnesium-manganese alloy. Grain growth does not occur until temperatures much in excess of 400° C. are reached. The tendency to grain growth, following critical cold work, although present in pure magnesium sheet, is much reduced by the presence of manganese in the alloy. It is noteworthy that the bending radius which can safely be put on this sheet alloy is approximately five times the thickness.

Extrusions

A large amount of tube is extruded in the 6% aluminium, 1% zinc, 0·2% manganese alloy (Elektron AZM), and this is used *inter alia* for the fabrication by welding of seats in aircraft. In addition, extruded rod and section is used for the manufacture of various aircraft components and as "stock" for forging. This alloy is most satisfactory for general purposes as far as the extrusion process is concerned, since it has a moderate aluminium content (the higher the aluminium content the slower the extrusion speed), while the presence of 1% zinc has a beneficial effect in assisting corrosion resistance. Extrusions in this alloy are not quite so strong as those with a higher aluminium content.

Forgings

Although a considerable number of forged components are made from extruded Elektron AZM, a number of quite

TABLE III.

COMMERCIAL MAGNESIUM WROUGHT ALLOYS USED DURING THE PAST 15 YEARS.

Alloy.		D.T.D. Specification.	Condition.	Average Tensile Properties.			American Equivalent.		A.S.T.M. Specification.	Average Tensile Properties.		
Elektron Designation.	Nominal Composition.			0·1% Proof Stress, Tons/sq. in.	Ultimate Stress, Tons/sq. in.	% Elongation on 2 in.	Trade Designations.	Nominal Composition.		0·2% Proof Stress, Tons/sq. in.	Ultimate Stress, Tons/sq. in.	% Elongation on 2 in.
AM.503	Mg + 1·5 Mn	142	Extr.	10·0	18·0	3·0	AM.3S Dow M	Mg + 1·2 Mn (max.)	B107—41T Alloy 11	13·4	18·7	7·0
AM.503	Mg + 1·5 Mn	118	Sheet	8·0	15·0	7·0	AM.3S Dow M	Mg + 1·2 Mn (max.)	B90—41T Alloy 11	7·1	14·3	16·0
AM.537*	Mg + 2·0 Mn + 0·5 Ce	—	Sheet	8·0	17·0	15·0	—	—	—	—	—	—
Zib*	Mg + 4 Zn	—	Sheet	7·0	14·0	10·0	—	—	—	—	—	—
AZ21*	Mg + 2 Al + 1 Zn	—	Sheet	6·5	13·0	10·0	—	—	—	—	—	—
AZ31*	Mg + 3 Al + 1 Zn	120A	Sheet	7·5	17·0	12·0	AM.53S Dow F	Mg + 3·3/4·7 Al	B90—41T Alloy 6	9·8	16·0	10·0
AZM	Mg + 6 Al + 1 Zn	259	Extr.	15·0	20·0	12·0	AM.57S Dow J	Mg + 5/8/7·2 Al + 0·4/1·0 Zn	B107—41T Alloy 8	13·0	17·8	16·0
AZM	Mg + 6 Al + 1 Zn	88B	Forging	8·0	15·0	5·0	—	—	—	—	—	—
AZM	Mg + 6 Al + 1 Zn	348	Tube	9·5	16·0	7·5	—	—	—	—	—	—
AZ.855	Mg + 8 Al + 0·4 Zn	88B	Forging	15·0	22·0	12·0	AM.58S	Mg + 7·8/9·2 Al + 0·2/0·8 Zn	B107—41T Alloy 9	15·6	21·0	12·0
V.1	Mg + 10 Al	259	Extr.	16·0	22·0	6·0	—	—	—	—	—	—
A.4	Mg + 4 Al	120A	Sheet	7·0	16·0	10·0	—	—	—	—	—	—

* These alloys are now seldom used. All the Al-containing alloys carry 0·15—0·40% Mn.

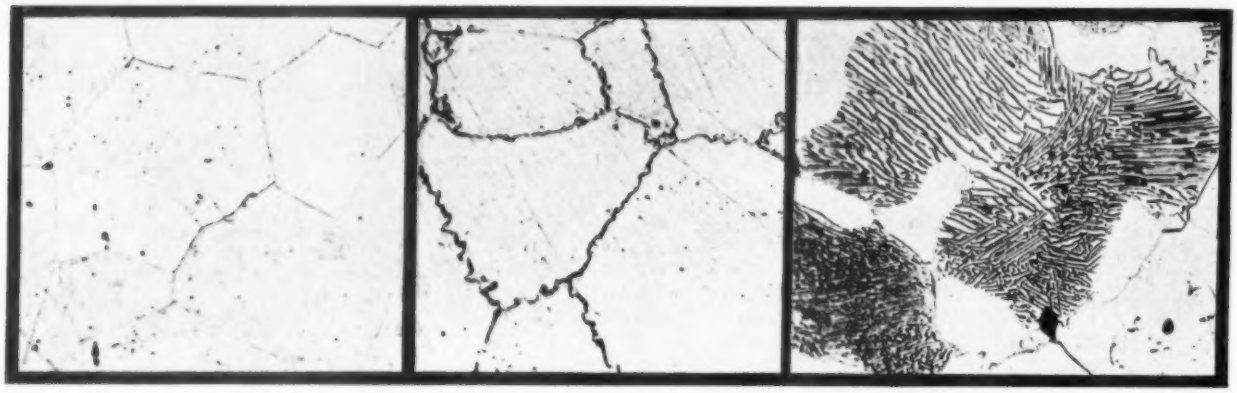


Fig. 7.—Elektron AZ.855 quenched after solution treatment, × 600.

Fig. 8.—Elektron AZ855 air-cooled after solution treatment. × 600.

Fig. 9.—Specially heat-treated Elektron A.8 showing "pearlite." × 450.

large forgings are produced in the alloy containing 8% aluminium, 0·4% zinc, 0·2% manganese; this is the alloy Elektron A.8. This alloy will not extrude rapidly, but it is sufficiently workable at temperature between 350° C. and 400° C. to be readily die-forged or pressed, and it is also capable of being hot-swaged. The forging temperatures required for small components are lower than those for corresponding parts in aluminium alloy. High pressures are required for forging large components, since it is important to ensure that the work shall penetrate the whole piece and shall not be confined to surface layers.

For the greatest uniformity of properties, and for the highest test figures, it may be desirable to forge extruded blanks; but it is readily possible, and in some cases even preferable, to forge cast blanks in this 8% aluminium alloy. The forging should be carried out relatively slowly, hydraulic pressure being the most satisfactory. The piece is also usually turned through 90° in successive operations, to ensure reasonably uniform distribution of the hot work and to avoid the production of excessive directionality. Forged components can be made in this alloy to resist large hydraulic pressures: for example, a hydraulic cylinder is now being made, the finished wall thickness of which is $\frac{3}{16}$ in., and the internal test pressure is 4,000 lb./sq. in.

Heat-treatment

The magnesium alloys containing aluminium and zinc are capable of being heat-treated, and undergo a marked improvement in mechanical properties as a result. For alloys containing only small amounts of zinc of the order of 0·5% or less, the solution treatment temperature is about 410° C., and 24 hours is usually the time which is applied commercially, although with alloys containing aluminium much above 8% solution is not complete under these conditions. Table IV indicates the improvement produced in two magnesium alloys by solution treatment, and by solution treatment and ageing.

The homogeneous state of solution treatment may be retained fairly readily in the aluminium containing alloys, and a liquid quenching is not essential unless the article is very large. It is interesting to observe, however, that an aluminium-containing alloy which is quenched after solution treatment shows a slightly different microstructure from that of a similarly solution-treated piece which has been air-cooled. The difference is illustrated in Figs. 7 and 8. The tortuous character of the grain boundary in the air-cooled specimen is connected with the incipient precipitation of the aluminium containing *beta* phase in the grain boundary. The solution-treated and aged alloy shows a "pearlitic" structure, as is illustrated in Fig. 9.

TABLE IV.

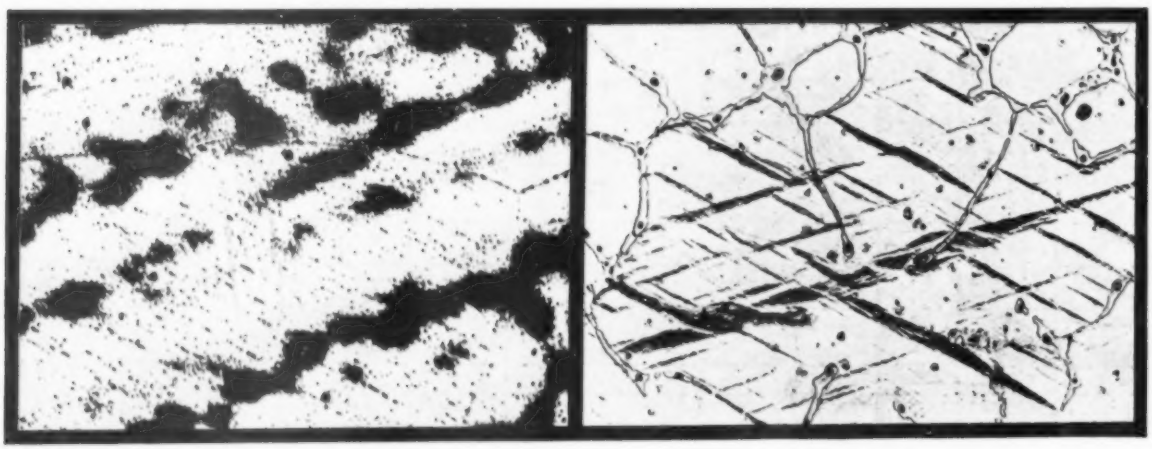
THE EFFECT OF HEAT-TREATMENT ON THE MECHANICAL PROPERTIES OF THE ELEKTRON ALLOYS A.8 AND AZ.91.

Elektron Alloy.	Nominal Composition.	Condition.	0·1% Proof Stress, Tons/sq. in.	Ultimate Stress, Tons/sq. in.	% Elongation on 2 in.
A.8	Mg + 8 Al + 0·4 Zn	As cast	5·3	10·5	3·0
A.8	" "	Soln. treated	5·1	17·0	12·5
AZ.91	Mg + 9·5 Al + 0·4 Zn	As cast	6·0	9·6	1·5
AZ.91	" "	Soln. treated	6·0	17·0	9·0
AZ.91	" "	Soln. treated and aged	7·5	17·0	1·5

All the Al-containing alloys carry 0·15—0·40% Mn.

Fig. 10.—Showing dendrite crossing grain boundaries in magnesium alloy casting. × 160.

Fig. 11.—Magnesium-cerium alloy showing true "grain" defined by strain markings. × 200.



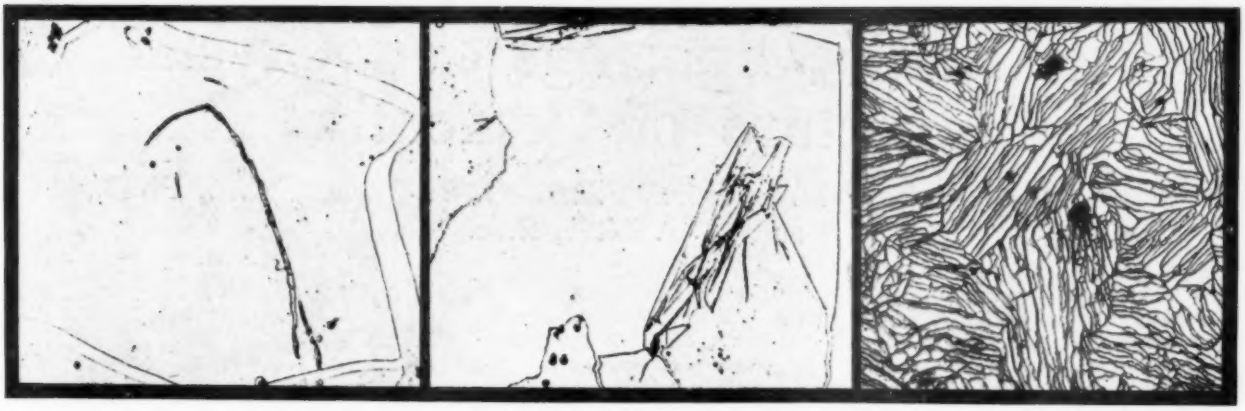


Fig. 12 —False structure caused by piece of an etching film. × 600.

Fig. 13.—Elektron AZ.855 quenched from 415° C. imperfectly etched, showing folding over of etching film. × 600.

Fig. 14.—Elektron AZ.855 quenched from 415° C., over-etched, showing rupture lines in etching film. × 600.

Here a low temperature ageing treatment has caused the precipitation, the commercial treatment being about 6 hours at 175° C. A more or less slowly cooled solution-treated piece, or even an "as-cast" piece, may show pearlitic precipitation.

This "self-ageing" has occurred during the time in which the metal was cooling, at temperatures below about 300° C; it has been found that although precipitation can occur, it is not the "pearlitic" type if the temperature is much above 300° C. In the range 300° C. to 360° C., precipitation occurs in the form of coalesced particles of *beta* phase. These different states of precipitation have an effect on the mechanical properties of the aluminium-containing alloys, although the differences are not outstanding.

It is evident that the change occurring on ageing solution-treated aluminium containing magnesium alloys is different from that occurring in the Dural type of aluminium-rich alloys. The change in mechanical properties in the magnesium alloys is directly related to the amount of pearlitic or visible *beta* phase precipitated, and the change in mechanical properties is not the direct effect of lattice distortion or small-scale slip interference, as is the case with Duralumin.

Micro-structure Features of Magnesium Alloys

The mechanical properties of magnesium are closely connected with the grain size, and an alloy will not be satisfactory if its grain size is large. The identity of a grain, even as shown under the microscope, is sometimes not clear. Fig. 10 shows a photomicrograph of a magnesium-tin alloy in which the dendrites cross the grain boundaries. The mechanism by which this effect is produced is not clear, although the phenomenon has been observed and reported in other non-ferrous metals. With some magnesium alloys it is necessary to apply strain to the specimen before the true individual grain is revealed. An example of this is illustrated in Fig. 11 ; without the indications afforded by the strain markings it would not be clear what were the limits of the grain in this case.

The aluminium-containing magnesium alloys are liable to etch-film formation, especially in the heat-treated condition. Fig. 12, for instance, shows that appears to be a grain boundary phase network, but it can be shown to be an etch film, and in Fig. 13 it is shown folded over locally. Fig. 14 shows an extreme case of etch-film formation with rupture lines in the film developed. It is noteworthy that "structures" such as these were, even in comparatively recent days of magnesium alloy develop- shown as genuine.

Surface Protection of Magnesium and its Alloys

Although many protection treatments for magnesium alloys have been reported, the most successful methods applied commercially are those of the type of a simple chemical dip, whether in cold or boiling solutions. In many applications it is desirable that the protective bath shall remove only negligibly small amounts of metal, and the solution treatment known as the R.A.E. half-hour bath fulfils this condition, as well as putting a satisfactory coat on the metal. This bath consists of 3·0% ammonium sulphate, 1·5% ammonium dichromate, 1·5% potassium dichromate, and 0·4% ammonia.

The degreased part is suspended in this boiling bath for 30 mins., washed in hot water, and dried. The film formed probably consists of mixed oxides of magnesium, chromium and manganese, and oxide of aluminium may be present when the alloy to be protected contains aluminium. If the casting is initially buffed, the finish, after treatment, is extremely good, and the corrosion protection is improved.

A second bath, known as the acid-chromate bath, consists of: 20–25 parts by volume of nitric acid, 100 parts by volume of cold water, 15% of the mixture by weight, potassium dichromate.

The magnesium parts to be protected are degreased, and then immersed for a short time, falling between 10 sec. and 1 min. They are then washed in clean water, and this is followed by a wash in hot water containing potassium bichromate. This bath produces a satisfactory film, is much more rapid in action than the R.A.E. bath, and can be operated at room temperature, but it has the slight disadvantage that a small amount of metal is dissolved, although, for the usual time of treatment, the amount of metal removed is very small, being of the order of 0·0005 in thickness.

It is noteworthy that the time during which the specimen is drained after the removal from the acid-dichromate bath, and before washing, has a very important bearing on the effectiveness of the film produced. In the case of Elektron AM.503 sheet, it has been found that a draining time of about 15 secs. seems to be optimum. If the specimen is drained for a much longer time—of the order of 1 min. or more—the film assumes a darker colour, and its protective value is somewhat decreased.

Severe galvanic action is likely to set in between magnesium and other metals in contact with it ; aluminium in particular, when coupled with magnesium, causes violent attack to be concentrated on the magnesium. For this reason it is essential in constructing assemblies of magnesium and aluminium alloys that the two metals should be well insulated from each other. It has been found that the aluminium-base alloys MG5 and MG7 have a much less harmful effect than has Duralumin. Stainless steels and cadmium plated, plain carbon steel, in electrolytic contact with magnesium, do not cause serious corrosion, although in practice it is necessary to ensure proper insulation.

The Lattice Spacings and Crystal Structure of Cementite*

By W. Hume-Rothery, M.A., D.Sc., F.R.S., G. V. Raynor, M.A., Ph.D., and A. T. Little, B.Sc.

Cementite was extracted by electrolytic attack from three steels of different compositions, and the lattice spacings were determined by Debye-Scherrer photographs in a 19-cm. camera. The values obtained were $a = 4{\cdot}5155$A, $b = 5{\cdot}0773$, $c = 6{\cdot}7265$ A at 25° C., and differed from those previously given by Lipson and Petch, whose specimen had partly decomposed during annealing. No difference could be detected between the lattice spacings of the three different samples, and the analysis of one of them confirmed the composition Fe_3C. The structure is discussed, and a possiblemechanism is suggested for the formation of cementite from martensite.

THE crystal structure and lattice spacing of cementite have recently been investigated by Lipson and Petch,[1] whose material was not actually extracted from a steel, but was prepared by passing carbon monoxide over ferric oxide at 550° C. The substance prepared in this way was used for structure analysis, but did not give sufficiently sharp lines for the accurate determination of lattice spacings, and these were determined after annealing the powder for 17 hours at 600° C. This annealing treatment resulted in the partial decomposition of the cementite, mostly in ferrite, but sometimes into another structure, which was probably Fe_2C. In view of the rather unusual method of preparation of the cementite used by Lipson and Petch, it appeared desirable to examine the cementite which is actually contained in steels, and the present paper describes the result of this investigation, which has shown that the cementite examined by Lipson and Petch had lattice spacings different from the constituent in steels.

The cementite examined in the present work was extracted from the steels, the compositions of which are given below, by electrolytic attack in hydrochloric acid, and was washed with distilled water, alcohol and ether. The resulting product did not give very satisfactory photographs, and it was therefore sieved in order to remove the very fine-grained material. The residue was gently broken up with a pin-head and again sieved, and the coarser portion used for the preparation of the X-ray specimen, which was mounted on a hair in the usual way with Canada balsam. The X-ray powder photographs were taken in a 19-cm. Debye-Scherrer camera, using filtered cobalt radiation. The steels examined were supplied by the Research Department, Woolwich, and had the following nominal compositions :—

Steel No. 1 :—1·07% carbon.
" No. 2 :—0·75% carbon, 0·25% titanium.
" No. 3 :—0·74% carbon, 0·99% titanium.

These steels were all of commercial purity, and contained approximately 0·5% of manganese, and 0·1% of silicon etc. The products extracted from steels Nos. 1 and 2 gave X-ray films with well-resolved sharp α-doublets, and in spite of a rather dark general background were excellent for visual measurement. The product from steel No. 3 gave a less satisfactory film, and is referred to only briefly below.

Analytical Results

The electrolytic product from steel No. 1 was examined analytically by both qualitative and quantitative methods. Only 0·3 grm. was available for analysis, and the composition was found to be :—

Iron, 92·95%. Total carbon, 7·05%. Free carbon, 0·53%.
Combined carbon = 7·05 — 0·53 = 6·52%.

The cementite thus contains 92·95 parts of iron to 6·52 parts of carbon, or 6·58% of carbon. With the small quantities used, the accuracy of the carbon determination is estimated to be of the order ± 0·05%, so that the figure for combined carbon is accurate to ± 0·1%, since two determinations are involved. In view of the small amount available, the agreement with the formula Fe_3C (6·68% carbon) is satisfactory, and the product examined may be regarded as possessing the usually accepted composition. Two independent tests failed to give any indication of the presence of manganese, and from the above figures it will be seen that, if the error in the carbon determination is 0·05%, the manganese content cannot in any case exceed 0·05%.

Lattice-spacing Measurements

With cobalt radiation, the outermost lines which are of importance may be summarised as follows: The outermost line consists of an overlap of the strong (432) and (145) lines, and the details of this overlap are considered later. The next two lines are the (136) and (251) lines, of which the $(136)\alpha_1$ and $(251)\alpha_2$ lines overlap, and the nature of this overlap is considered later. The $(136)\alpha_2$ and $(251)\alpha_1$ lines are quite clear of the overlap, and can be used for accurate measurement.

The next suitable line is the (334), of which the α_1 component is overlapped by the $(423)\alpha_2$ line and is useless. The $(334)\alpha_2$ line lies in between the α_1 and α_2 lines of the (226) reflection, but this is so faint a line that in the 19-cm. camera the $(334)\alpha_2$ line can be used to a reasonable degree of accuracy. The next satisfactory line is the (107) line, of which the α_1 component is clear and satisfactory, although the α_2 component is overlapped by the $(430)\alpha_1$ line. The next two strong lines are the (152) and (235) lines, both of which are clear and satisfactory.

For the determination of the axial ratios we have therefore the following lines :—

$136\alpha_2$	$334\alpha_2$	$152\alpha_2$ and α_1
$251\alpha_1$	$107\alpha_1$	$235\alpha_2$ and α_1

These lines range from 77·79° to 69·29°, and their combination is very fortunate, since the high values of k and l enable the one axial ratio to be determined accurately. The structure of cementite is orthorhombic, and the equation for the b side of the unit cell may be written in the form :—

$$b = d\sqrt{\frac{h^2}{N^2} + k^2 + \frac{l^2}{M^2}}$$

where d is the interplanar spacing, and $N = a/b$, and $M = c/b$.

If, therefore, the axial ratios are chosen correctly, the values of b calculated for the different lines from this expression should lie on one straight line when plotted

* Iron and Steel Institute paper. March, 1942. (Advance copy.)

1 H. Lipson and N. J. Petch. *Journal of the Iron and Steel Institute* 1940, No. II., p. 95P.
2 A. J. Bradley and A. H. Jay: *Proceedings of the Physical Society*, 1932, vol. 44, p. 563.

against $\cos^2\theta$, as in the method of Bradley and Jay.[2] Using the axial ratios given by Lipson and Petch, the different points did not lie on one straight line, but, on adjusting the axial ratios to the values given in Table I, a satisfactory straight line was obtained, and the high values of k and l in the (136), (251) and (107) lines mean that the axial ratio $M = c/b$ is established to a high degree of accuracy. It was also noticed that the revised axial ratios given in Table I required an almost exact overlap of the $(136)\alpha_1$ and the $(251)\alpha_2$ lines. Calculation shows that the peaks of these lines should differ by about 0·08 mm. on the film of the 19-cm. camera. This overlap line was then examined in a Hilger photo-electric microphotometer and was found to consist of a sharp, well-defined line with a horizontal top, of width between 0·1 and 0·2 mm. No great significance attaches to the actual value, since the films were exposed for visual examination, and were thus over-exposed for the best results from the photometer. The photometer record showed clearly that there was an almost but not quite exact overlap of the $(136)\alpha_1$ and $(251)\alpha_2$ lines, whereas the axial ratios of Lipson and Petch would require a separation about ten times as great. This furnishes additional proof of the correctness of the choice of the axial ratio $M = c/b$, but on account of the low values of h in the (136) and (251) lines it is not a sensitive test for the accuracy of the second axial ratio $N = a/b$. It was, however, found that the axial ratios given in Table I were such that the values of $\sin\theta$ for the (432) and (145) reflections were identical to within 1 part in 50,000. The high-angle doublet on the film was then examined visually and by means of the photometer, and was found to have the characteristics of a single reflection.* This confirms that the second axial ratio, $N = a/b$, has been chosen correctly, since the values of h are very different in the two lines, and the point for the single (432,145) line lay on a satisfactory Bradley-Jay $\cos^2\theta$ straight line through the remaining points.

We have thus two independent checks that the correct axial ratios were deduced by the adjustment to obtain a single straight-line Bradley-Jay $\cos^2\theta$ plot. Having established the axial ratio in this way, the b lattice spacing was readily determined by the usual $\cos^2\theta$ method: the outermost (432,145) line lies at approximately 80·69°, and has a satisfactorily high angle.

TABLE I.
LATTICE CONSTANTS OF CEMENTITE.

	Present Work.		Lipson and Petch.
a	4·5155	..	4·5144
b	5·0773	..	5·0787
c	6·7265	..	6·7297
a/b	0·88934	..	0·888888
c/b	1·32481	..	1·32508

This work was carried out on the cementite extracted from steel No. 1, and identical results were obtained from teel No. 2. As explained previously, the film of the cementite from steel No. 3 was very much less satisfactory. For four of the outermost six lines of this film, the angles of reflection agreed within 0·02° with those of the first two films, and for the remaining two lines the differences were 0·04° and 0·08°, respectively. This film was not accepted as suitable for accurate measurement, but the agreement of the angles is sufficiently close to show that no appreciable difference exists between the lattice spacing of this and the other two samples of cementite, and to confirm the present results, in contrast with those of Lipson and Petch. It may therefore be reasonably claimed that the present figures are charactertistic of the cementite found in three quite different steels. Correspondence with Dr. Lipson has shown that the lattice spacings that he obtained were undoubtedly correct for the sample which he examined, and the difference between the two investigations is presumably the result of the partial decomposition during the annealing of the cementite used by Lipson and Petch. If it is accepted that a change in composition is accompanied by a change in lattice spacing, the present results may be taken to indicate that the cementite in these steels is of constant composition.

* The axial ratios of Lipson and Petch require a considerable separation of the two lines.

The Martensite Cementite Transformation

As pointed out by Lipson and Petch, the crystal structure of cementite appears to bear little relation to the structures of either α- or γ-iron, and any simple connection between these structures is difficult to trace. It is well known that cementite is formed from martensite at considerably lower temperatures than from austenite, and this fact suggests that the relative positions of the iron and carbon atoms in martensite are specially favourable for this transformation, and so enable it to take place with comparatively small atomic movements at a low temperature. The following description is intended to show that the structure of cementite can in fact be derived from that of martensite by a comparatively simple shearing process, and it is tentatively suggested that this may explain the ready formation of cementite from martensite.

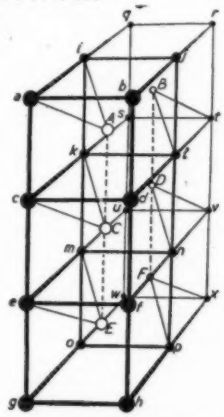


Fig. 1.—Body-centred cubic structure (martensite).

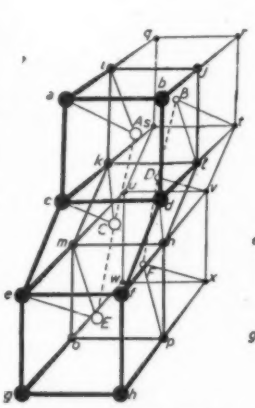


Fig. 2.—Shear displacement of half length of cube side.

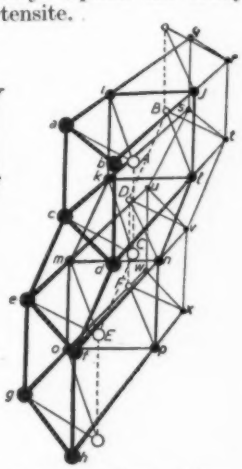


Fig. 3.—Final adjustment produces cementite structure.

Figs. 1 to 3.—Formation of cementite structure from martensite structure by shear. In these diagrams only the positions of the iron atoms are shown. The atoms at the corners of the body-centred cubes are represented by full circles, the size of which diminishes as the atoms recede. The atoms at the centres of the cubes are represented by open circles, the size of which again diminishes as the atoms recede. Individual atoms are identified by means of letters.

In austenite the carbon atoms are almost certainly placed at the centres of the unit cubes, and are thus surrounded by six iron atoms at the corners of an octahedron. It if is accepted that, after the formation of martensite, the martensite (110) plane and [111] direction are parallel to the (111) and [110] direction of the parent austenite, the most probable positions of the carbon atoms are at the centres of the square faces of the unit of the body-centred tetragonal cell. Martensite is tetragonal, but the axial ratio is only 1·03, so that the structure is very nearly cubic. For simplicity in description the structure of martensite will be drawn as that of a body-centred cube, since this enables the changes to be grasped more readily and does not invalidate the argument. Fig. 1 shows six unit cells of a body-centred cubic structure, and the carbon atoms are to be thought of as occupying the centres of square faces such as *aikc*, *ckme*, etc. In the actual tetragonal structure those faces remain square, and the structure is to be regarded as that of Fig. 1 stretched slightly in the direction of *ab*.

If, now, a simple shear takes place in the central layer of cubes, such that the top and bottom layers are displaced relatively to each other by an amount of the order of $\frac{1}{2}a$, where a is the length of a side of the unit cube, the stage

shown in Fig. 2 is reached. The displacement has taken place in the [100] direction. This structure contained several features which show similarity with the structure of cementite as regards the general atomic configuration, and may be regarded as a distorted form of the cementite structure. Comparatively minor adjustments of angles and distances enable the cementite structure to be formed, and the final structure is shown in Fig. 3.

The essential changes in this final adjustment may be conveniently summarised as follows:

(*a*) The carbon atoms take up positions which are alternately a little outside or inside the planes of faces such as *aikc*, *ckme*, etc. In the diagram, carbon atoms should be placed outside the planes *aikc*, *emog*, *jrtl* and *nvxp*, and inside the planes *ckme* and *ltvn*. This displacement of carbon atoms allows a contraction of the faces concerned, so that, for instance, the distances *ai* and *ac* contract while the face remains approximately square. It is clear that this must lead to an expansion of distances such as *lj* and *iq*, while the angle between the planes which were the vertical faces of the parent cubes is no longer a right angle, but greater. The distances such as *ab* also contract.

(*b*) At the same time the distance of the central atoms of the sheared middle section of the structure from corners such as *c* and *k*, or *n* and *v*, increases in such a way that these atoms (e.g., *C* and *D*) lie either vertically below the central atom of the unit above or vertically above the central atom of the unit below. Thus, atom *C* falls vertically below atom *A*, and atom *D* vertically above atom *F*.

(*c*) It must also be assumed that the distances *iA*, *aA*, *bC*, *cC*, etc., also increase from a value close to that characteristic of the body-centred cube (approximately 2·48 A. for α-iron) to the value of 2·68 A. characteristic of cementite.

It is possible that a mechanism of this nature is involved in the ready formation of cementite from martensite. It is particularly suggestive that a simple shear of alternate body-centred cubes gives a structure the general configuration of which resembles that of cementite. Contraction of the sides of faces from which carbon atoms are expelled, followed by approximate equalisation of the distances between neighbouring iron atoms, gives a structure of exactly the same type as that of cementite. Although the authors have had no opportunity of testing it experimentally they have thought it of interest to advance this hypothesis, which appears to explain a very characteristic change.

Die-Block Heat-Treatment

By Bernard Thomas, F.Inst.P., F.Inst.F.

As a result of recent tests on nickel-chromium-molybdenum grade 5 die steel, to determine the most suitable heat-treatment to apply, it is suggested that die-blocks of this quality are quenched in oil from 830° C. and tempered at 640° C. and quenched in water, soaking time to be determined by the size of block under treatment.

BOMBS, more bombs—and planes from which to drop them. And neither could be produced without drop-forgings—and drop-forgings are produced in die-blocks. And so by stages we arrive at the conclusion that die-blocks are playing a vitally important part in the conflict raging. Are users satisfied that they are obtaining the best service possible from these tools? The object of these remarks is to make known the results of certain tests carried out recently to determine the best treatment to apply to what is probably the most used steel for the purpose in question.

Ignoring special steels of the hot- and cold-die types, the chief favourite for general stamping appears to be the nickel-chromium-molybdenum No. 5 Grade as covered by Specification B.S.S. 224–1938. The composition of such is given in Table I.

TABLE I.
COMPOSITION OF N.C.M. GRADE 5 STEEL.

Carbon	0·50–0·60%
Silicon	0·30% max.
Manganese	0·50–0·80%
Sulphur	0·05% max.
Phosphorus	0·05% max.
Nickel	1·25–1·75%
Chromium	0·50–0·80%
Molybdenum	0·20–0·40%

Some forgers purchase the blocks in the softened condition and harden and temper after sinking, holding the view that a saving is thus made in die-sinking costs by reason of cutting a soft, instead of a heat-treated material. This can only hold good for one sinking, however, and the purchaser must then either proceed to sink on the now treated face, or alternatively soften, re-sink and re-harden. Production pressure may not always allow of the latter procedure, in which case the initial advantage is nullified after the first sinking has worn.

Other users may prefer to obtain their blocks already hardened and tempered by the supplier, the treatment applied being either as laid down by the specification ruling, or alternatively to yield Brinell hardness figures as requested by the customer. In the former case the blocks are quenched from 800°–830° C. and tempered at a temperature not exceeding 650° C.

Unfortunately, not only does mass effect play an important part in the figures yielded, but die-block suppliers invariably lack knowledge of the type of material to be stamped in the die; consequently many dies may enter service not in the condition best suited for the job involved. Such circumstances result in the inevitable loss of production we all try to avoid.

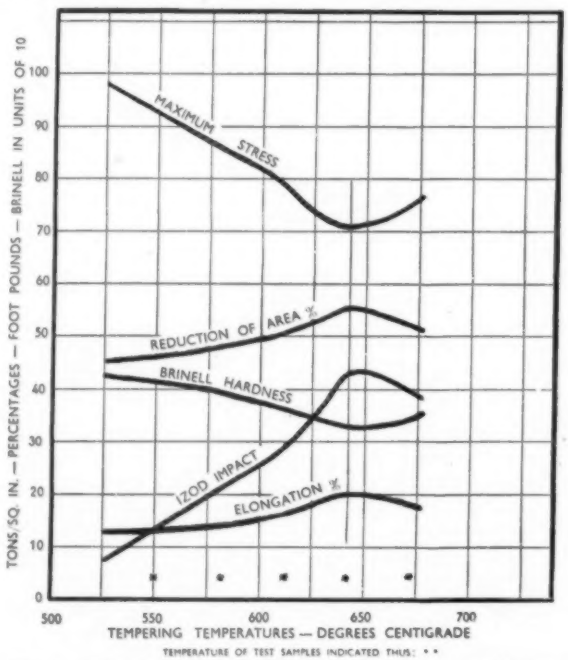


Fig. 1.—Physical results obtained from a series of heat-treatment tests.

It was for reasons outlined above that the tests under description were carried out, in the belief that the best

average results must surely be obtained from blocks having the highest combination of physical properties and subsequent results have fully substantiated such reasoning. There have been, of course, exceptions to this, like any other rule, but at least a tangible treatment has been determined.

A number of test-pieces were prepared from NCM Grade 5 steel measuring 1¼ in. in diameter, care being taken to ensure that direction of grain ran longitudinally. These pieces were all quenched in oil from 830° C. after a soak of 1 hour. Tempering was carried out at rising increments of 30° C., commencing from 550° C. and ending at 670° C. In Table II is depicted the physical results obtained, while Fig. 1 shows similar matter in graph form for ease of assimilation.

TABLE II.

PHYSICAL RESULTS ON TREATED TEST-PIECES OF NCM GRADE 5 STEEL.

Test Mark.	Tempering Temperature, ° C.	Ult. Stress, Tons per Sq. in.	Elongation, %.	Red. of Area, %.	Ave. Izod Value.	Brinell Impression, M/m.	Brinell Hardness 10 m/m, 3,000 Kg.
A	550	91·8	13	46	13	3·00	415
B	580	85·2	14	48	20·7	3·10	388
C	610	80·3	16	50·5	28·3	3·20	363
D	640	70·8	20	55	43	3·40	321
E	670	75·1	18	52	39·7	3·30	341

Probably the graph in Fig. 1 shows most clearly the peak properties obtainable by tempering at 640° C. Not only is the strength still high at roughly 71 tons per sq. in., but the Izod impact value shows an increase of 50 per cent. over the next lower tempering temperature.

It may occasion some surprise to note the increase in hardness obtained by tempering at 670° C., coupled with the falling-off in the Izod and ductility values. This is, however, an occurrence in line with the treatment of other nickel-chrome type steels, such as 4S11 and S65, where a temper in excess of 640° C. results only in an *increase* of hardness and not decrease, due presumably to the air-hardening propensities. One does not normally expect, however, the lower nickel content of the die-block quality to effect such changes, but the evidence lies in the test-piece.

In summary, therefore, it is suggested that die-blocks of this particular quality are quenched from 830° C. in oil, and tempered at 640° C. and quenched in water, soaking time in each case to be determined by size of block, with 1 hour per inch of thickness from side to centre as a rough guide—e.g., an 8 in. × 8 in. × 8 in. block would be 4 in. from side to centre, and 4 hours would be a reasonable soaking time for both hardening and tempering. Actually, in the experience of the writer, if the *tempering* soak is double that which appears necessary, a greatly improved life is resultant, and some of the finest results have been obtained from blocks tempered 12–14 hours at 640° C.

Another point is the necessity for early transference from the oil after quenching to the tempering equipment. This should be done while the block is at a temperature of 350°–400° C., as the finer condition of the martensitic structure so obtained, appears to lend itself more readily to the formation of the lower transitional constituents, with consequent improvement in physical properties.

Solvent Economy in Trichlorethylene Degreasing Plants

The solvent degreasing of metal components has grown in importance in recent years, but with the trend in practice to degrease work in the same steady flow as with preceding operations, instead of in batches, trichlorethylene is likely to be used in an economical manner, unless reasonable precautions are taken. This aspect of degreasing was discussed by Mr. W. F. Jesson at a recent meeting of the Electrodepositors' Technical Society, the salient features of which are given in this article.*

WHEN the advantages of solvent degreasing were attracting the attention of the metal trade 14 years ago, though the chemical and physical properties of trichlorethylene had long been known, there was much to be learned concerning its behaviour in commercial degreasing plants under workshop conditions. As is frequently experienced, experimental results could not always be reproduced in factories, and it became clear that the design of degreasing plants had to be considered more from the point of view of the manufacturing requirements than of theoretical considerations. The latter suggested totally enclosed plants to degrease work in batches, while the former needed a plant which would degrease work in a steady flow.

The first plants were enclosed and the lid carried the cooling coil which prevented the vapour escaping. Whether the load of work was large or small, the same degreasing cycle had to be carried out—viz., the vapour produced by the boiling solvent in the sump of the plant condensed on the lower part of the work above it, and this condensate dissolved the grease so that both dripped off together. In condensing the vapour the lower part of the work became heated to the same temperature as the vapour, and this allowed the latter to rise gradually up through the work until it reached the cooling coil in the lid. The flow of water through the lid-cooling coil prevented the vapour from rising further by condensing it to a rain of solvent which fell upon the work beneath, thus adding a washing effect. When the work was completely degreased the cooling water was diverted from the lid coil to another cooling coil placed above the sump, but just beneath the work; by this means the hot vapour above the coil was condensed, and the vapour coming from the sump was maintained at the level of this coil, so that the lid could be opened and the load removed.

The weakness of this design was that each time the vapour was lowered air was drawn in to take its place. This air was warmed by the sides of the tank and was thus enabled to hold a higher concentration of solvent than when it was cool. When the vapour was raised to degrease the next load of work this solvent impregnated air was displaced from the tank and dispersed. If the tank was completely filled with work at each degreasing cycle the quantity degreased per gallon of solvent could be satisfactory, but if the work was treated in small batches the solvent consumption was excessive.

The most economical design of plant working on the batch principle is one in which the work is degreased in vapour or boiling liquor in a closed vessel, and is dried off before being removed into the atmosphere, by means of circulated air alternately heated before passing through the work to drive off entrained solvent and cooled afterwards to condense and recover the solvent. Such plants are at their best in degreasing substantial loads of small parts, and in this they are remarkably economical, but they suffer from the disadvantage of rather complicated and slow operation compared with the simple open-topped type.

Should the importance of degreasing work in a steady flow be greater than that obtained by the highest economy of solvent, then the open-top type of plant is preferable. A compromise between the economy of the batch type and the simpler working of the open-top type is obtainable by

* *Jour. Electrodepos. Tech. Soc.*, 1942, **17**, 34–44.

equipping the latter with a conveyer to suit the class of work to be degreased : the conveyer mechanism passes into and out of hooded turrets at the ends of the plant, the rest of it being completely enclosed. Where the conditions are suitable such plants can show marked solvent economy while continuously delivering degreased work at a uniform speed.

The majority of degreasing plants in use to-day are of the open-top type, having constant vapour level. These plants are broadly classified into the vapour type and the liquor type, but for the present consideration of solvent consumption, what applies to one applies also to the other, since, in each case, the work is lifted out of the plant through a vapour zone. The solvent consumed in these degreasing plants is part of that which dries off from the degreased work, and that which diffuses into the air from the vapour surface. Considering the former, the most important factor in economy is to withdraw the work slowly and steadily from the vapour surface so that, as it dries off, a layer of concentrated trichlorethylene in air is formed and is subjected to the minimum of disturbance just above the vapour surface. Such a layer is heavier than air, it is largely condensible by the cooling coils, and it checks diffusion ; its maintenance is to a large extent dependent on freedom from disturbance : this in turn is affected by the rate of withdrawal of the work, so the latter has to be a compromise between the theoretical optimum and the practical requirements of work output.

A conveyer speed of 10 ft. per min. is usually aimed at in mechanised plants, but in hand-operated plants the rate of lifting the work while it is in the drying-off zone within the plant should approximate to 5 ft. per min. The operators, however, are frequently paid on the quantity of work degreased, and are seldom, if ever, paid any bonus upon solvent saved, and naturally they do not all exercise the care necessary to enable most of the solvent to dry off near the vapour surface.

It should be noted that just as hasty raising of the work from the vapour lifts the latter up into the air. so hasty lowering of the work into the vapour causes air to be carried down into the latter and makes that mixture of solvent and air which is the prime cause of waste. This mixing of solvent and air is more difficult to check where the shape of the work prevents drying off as rapidly as occurs from a flat surface held vertically, for example, masses of small screws tend to hold vapour as a sponge holds water, while hollow work can cup solvent in liquor or vapour form and cause excessive waste unless special work containers are used to meet these special needs.

The foregoing causes of solvent waste depend largely on the operator, the shape of the work, and the way in which it is handled. There is, however, a loss of solvent from an open-topped plant, while no work at all is being degreased : this loss results from the mixing of solvent vapour and air according to the law relating to the diffusion of gases. This refers to the molecular activity between the contacting surfaces of gases so that each penetrates the other ; such diffusion is most rapid between pure air and concentrated vapour. The high concentration of solvent vapour in air which results becomes more dilute as it extends and though this dilute zone checks the rate of diffusion the latter cannot be completely stopped from an open vessel. It is the prevention of diffusion from enclosed degreasing plants and the protection of the drying-off zone from air movement in the shop which confers on them their greater economy in solvent consumption.

A compromise between the economy of the enclosed type of plant and the handy working of the open-top type can be obtained by making the fullest possible use of the lids with which the latter is provided. If the plant is closed by a lid during idling periods the escape loss is small : if the lid is left open the loss is appreciably greater than when the surrounding air is still, but draughts can make the loss serious. Considerations of solvent economy, therefore, demand still air over the plant, while the ordinary requirements of health necessitate good ventilation. These requirements are not contradictory ; the open top of the plant should certainly be protected from air moving rapidly across it, but it is important that there should be plenty of head room so that any solvent getting into the atmosphere can pass freely away above the operator's head.

Other points in the design and installation of plants which affect solvent consumption are :—

1. There must be the correct amount of freeboard or height to the top of the plant from the vapour level. Experiment and experience have shown that there is an optimum figure for this related to the length and width of the plant.

2. The cooling surface must be sufficient to hold the vapour down to the correct level when the maximum heat is being applied, and no work is being degreased. The depth of the cooling coil should be sufficient to condense the rich Tri-air just above the vapour level, which is mostly produced by the drying off.

3. Wasteful air currents are sometimes caused by a hot flue pipe rising close to the plant ; these can be checked by fixing a vertical screen on the plant side next to the flue, or by using a hinged lid for that purpose.

If the foregoing points have been attended to properly the amount of solvent loss should be small in comparison with that consumed in the actual degreasing of work.

The solvent consumed in the degreasing of metal is governed primarily by the area of the surface degreased and its shape—i.e., the area from which solvent dries off into Tri/air insufficiently concentrated to be condensed on the cooling coil. The greatest area of metal which can be degreased per unit of solvent in an open-topped plant is in the form of flat sheets. A figure as high as from 4,000 to 5,000 sq. ft. per gal. has been obtained regularly in workshop conditions, and this includes any losses in distillation and cleaning out of the plant. But were these steel sheets to be pressed up into irregular shapes, such as parts of motor-car bodies, and to be degreased again, then the solvent consumption would be considerably increased owing to greater disturbance of the vapour as the work is lifted out of it and also to the tendency for irregular surfaces to carry the vapour higher up in the plant before drying off ; this would increase the Tri/air concentration in the upper part of the plant whence it would be more readily dispersed. Thus, it is not possible to quote a definite surface area for all classes of sheet-metal work which can be degreased by 1 gal. of solvent.

A more common measure of the quantity of work treated is in the terms of weight. Another measure of solvent consumption is to check the amount used per week. It does not attempt to assess surface or weight, and it therefore appears to be a rough and ready method, but, properly used, it is probably the best basis on which to organise that workshop control of solvent consumption which is essential to economy.

Broo-Zinc Sand and Die-casting alloy

The composition of Broo-Zinc is in the range of 58–62% zinc, 35–38% aluminium, and 2–4% copper ; in addition, the alloy contains certain well-specified purifying agents which not only improve the physical properties, but increase resistance to corrosion and limit intercrystalline growth. Although originally developed for sand- and die-castings, the alloy is now available in the form of rods and extruded sections.

Data contained in a recent booklet indicate that Broo-Zinc has qualities which make it an excellent substitute for many aluminium alloys. It has a comparative low specific gravity, 4·5 to 4·8, good dimensional stability, resistance to corrosion, and these factors, combined with good casting properties, have contributed to its increasing use for many sand- and die-castings. Useful information regarding the properties of this alloy is given in the booklet available from the manufacturers, Brookside Metal Co., Ltd., Astra Works, Honeypot Lane, Stanmore, Middlesex.

The Developments and Characteristics of "Substitute 66" High-Speed Steels

By A. Linley, B.Met.

Chief of Research, Darwins, Ltd., Sheffield.

The possibility of using molybdenum in place of or additional to the other metals—tungsten, chromium, vanadium and cobalt—in high-speed steel has been known for many years. As a result of continuous research work, molybdenum high-speed steels are now widely used, and in this article special reference is made to the steel containing 5 to 6% molybdenum and 5·5 to 6·5% tungsten, which is rapidly becoming popular. It can be accepted with confidence as comparable in many respects with tungsten high-speed steel.

BECAUSE a restriction in the use of tungsten has become necessary, two alternative steels have been introduced in which tungsten is in some degree replaced by the metal molybdenum. These steels have been designated "Substitute 66" and "Substitute 94." These notes deal only with the former material because it is on this that the writer has the greater knowledge and experience.

At the outset, let it be said that the use of the title "substitute" in reference to these molybdenum high-speed steels and their replacement of tungsten steels has been to some extent misleading. "Substitute" often conveys the same atmosphere as *ersatz*—a word which has come to signify something which replaces, but is not quite so good as an original material. "Substitute" steels, however, are certainly not *ersatz* materials. They are a natural result of research and experiment and their development follows the normal course of metallurgical progress. The fact that they have arrived with some suddenness (causing slight apprehension in some quarters) is due solely to an emergency which war-time happenings have created.

Artificial rubber has been found to possess certain properties which cannot be obtained from natural rubber; in like manner, molybdenum high-speed steels, by virtue of their efficiency, will be found to meet certain conditions which will remain their prerogative even when tungsten supplies are again abundant.

Much stress has been laid on the replacement of 18% tungsten high-speed steels by these alternative materials, and this has led to the inference that molybdenum high-speed steels cannot approach the cobalt high-speed steels in performance. Such belief is erroneous. In some operations molybdenum high-speed steels have shown results equal to those of the cobalt types. At the same time, it is not claimed that molybdenum high-speed steels are a type of "universal" material for all machine operations. But neither is any type of high-speed steel!

The History of Molybdenum High-speed Steels

The following short history is given to stress the fact that molybdenum high-speed steels are not products developed merely to meet an emergency, and that very little knowledge of them, and probably less experience, is available.

Early work in the development of the steels can be traced to the beginning of the century; during the lifetime of the inventors of modern high-speed steels—Taylor and White—the ability of molybdenum successfully to replace tungsten was known. In 1907 Taylor said that he and his co-inventor knew that in their patent "one part of molybdenum will produce approximately the same effect as two parts of tungsten."

In common with other workers, Taylor and White experimented with the use of molybdenum for high-speed steels. Taylor reported, however, that where the molybdenum content was above 2% the steels showed a tendency to irregularity in performance; above 4% the irregularity was definite. The irregularity, Taylor suggests, was due to loss of molybdenum in the form of molybdenum oxide or to surface decarburization.[1] French and Digges[2] again refer to the loss of molybdenum, and mention the use of cobalt to prevent this. They also make mention of fumes during forging and rolling. Grossman and Bain,[3] in 1931, confirmed Taylor's statement regarding the volatility of molybdenum oxide (MoO_3) and also stated that molybdenum diffuses rapidly, aiding the loss of the actual metal. They repeated Taylor's further contention that when the molybdenum content of a high-speed steel is restricted to 1 to 2% the effect of the loss of molybdenum in processing is not serious.

There had been replacement of tungsten by molybdenum wholly and partially, but it was apparent that molybdenum high-speed steels were not a commercial success, and the difficulties of processing and soft skin pointed out by Taylor still remained to puzzle those engaged in research and manipulation.

During 1927-1930 a marked advance took place in the development of molybdenum high-speed steels when the United States Arsenal at Watertown began a series of researches which were made public in 1930 by Ritchie. After initial experiments with the partial substitution of tungsten by molybdenum the Arsenal workers proceeded to complete substitution and eventually it was considered that the best all-round molybdenum high-speed steel had the following composition:—C, 0·80; Mn, 0·25; Si, 0·25; Cr, 3·5; Mo, 9·5; V, 1·25. This steel was considered to be 90% as efficient as 18% tungsten high-speed steel, and for some special purposes to be even more efficient.

It was also considered that the soft skin which had often been noticed on molybdenum steels and thought to be due to molybdenum loss, was, in fact, a result of decarburisation. The use of borax to prevent this was suggested. It was also found that the temperatures for forging the steels could be too high and should not be as high as those for the original high-speed steels. The maximum temperature suggested was 1,150° C. (preferably 1,100° C.), and the minimum temperature 950° C.—a startling similarity to the procedure advised for modern molybdenum high-speed steels. The report prepared by the Arsenal also stated that the 9·5% molybdenum steels, free from tungsten, forged more easily than 18% tungsten high-speed steel. Further, a valuable conclusion from these experiments was that hardening temperatures were below those for 18% tungsten steels. The Arsenal also advised the use of salt baths.

From the foregoing somewhat sketchy summary of the major points of the report issued by the United States Arsenal, it will be observed that many of the recommendations to-day are not new, and that, in point of fact, they originated as far back as 1930.

The work undertaken by the Arsenal revived interest in molybdenum high-speed steels, and J. V. Emmons, of the Cleveland Twist Drill Co., carried out tests on drills made

1 F. W. Taylor. "The Art of Cutting Metals." Trans. Amer. Soc. Mech. Eng., vol. 28, 1907.
2 French and Digges. Trans. Amer. Soc. "Steel Treating." 1925, vol. 8.
3 Grossman and Bain. "High-speed Steel." 1931.

from the steels. Emmons found that they were not as efficient as 18% tungsten steels. He therefore added a small amount of tungsten to a steel similar to the Watertown steel, and took out patents in the U.S.A. Emmons called the steel "Mo-Max," and its composition was :—C, 0·80 ; Cr, 3·5 ; Mo, 9·0 ; W, 1·5 ; V, 1·25.

This steel was taken up by other manufacturers under licence, and by 1938 there were at least nine makers.

Substantial tonnages of the steel have been made (and continue to be made), but in all its processes—forging, rolling, hardening, etc.—borax coatings have to be used to prevent the formation of the soft skin which is still the bugbear of the Emmons' steel. Unceasing attempts have been made to overcome this disadvantage, and to make molybdenum high-speed steels as easy to manipulate as the old standard high-speed steels, copper and boron additions were made without showing much improvement. Eventually, the development of molybdenum high-speed steels took a logical trend. It was admitted that molybdenum tends to cause soft skin ; therefore, it was argued, why not decrease the tendency by decreasing the molybdenum content to a point where soft skin does not occur and balance the loss of efficiency caused by the lower molybdenum by an increase in tungsten ? This was done,

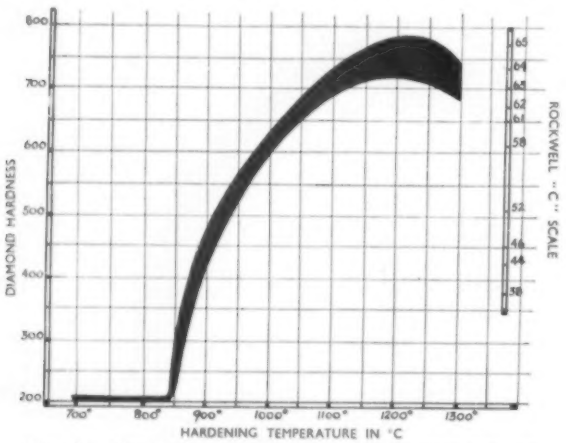


Fig. 1.—Hardening curve of 6/6 type substitute as quenched and untempered.

and the result was quite successful—hence the development of the "66" type of steel. For this credit may be claimed by the Allegheny-Ludlum Steel Corporation, the Crucible Steel Company, and the Climax Molybdenum Co. Already the "66" type of steels—i.e., "Substitute 66"—and their variations are proving to be the most popular molybdenum high-speed steels in the U.S.A., and this year will probably outstrip all the other varieties.

Following is a brief summary of the advantages of "Substitute 66" steel as compared with 18% tungsten steel :—

(*a*) Forging and rolling is easier and can be carried out at a lower temperature.
(*b*) In the annealed condition the material is easier to machine.
(*c*) The heat-treatment temperature is lower.
(*d*) The specific gravity is lower, therefore more steel is obtained per lb. weight.

The Manipulation of "Substitute 66" Steel

Forging

This operation is one of the most important operations from a metallurgical point of view, and demands the closest attention on the score of refinement of structure. Tungsten high-speed steels are notorious for the persistence of their ingot structure, which has to be altered as much as possible if directional brittleness in tools is to be eliminated.

Experience in the U.S.A. has shown that molybdenum high-speed steels are less subject to persistency of structure. Thus for equal forging (i.e., reduction) the molybdenum steels will yield better metallurgical structures. That is to say, there is less "ingotism" and fewer carbide stringers. Since the last two conditions are a common cause of brittleness in high-speed steel, the fact that they are present in less degree in molybdenum high-speed steels may be the reason why tools in these steels can be used at slightly greater hardness than equivalent tools in tungsten steels.

The forging temperature for "Substitute 66" steel is lower than for standard tungsten high-speed steels. Forging should be carried out at 1,040°–1,150° C. maximum, and be discontinued when the temperature falls to 950° C. For this type of steel no special atmosphere furnace or borax coating is necessary, though, obviously, the better the atmosphere in the furnace, the less the scaling which will occur.

The finish forging temperature is above the upper change point ; therefore hardening will take place if the steel is allowed to cool from forging heat. Very often a tool or bar is forged at one end only, and this results in a graduation of temperature, an occurrence likely to lead to severe internal strain on cooling. The method which should be adopted is the same as that for ordinary high-speed steels. The cooling should be done as slowly as possible in ashes, kieselguhr, or other insulating material.

Annealing

Softness is essential for ordinary machining operations on high-speed steel tools. Even where such parts as tool bits are ground only, rough grinding is done best in the soft condition, leaving the minimum amount of grinding to be done in the fully hardened condition.

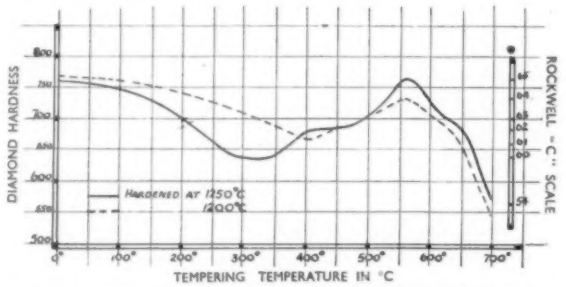


Fig. 2.—Tempering of 6/6 type molybdenum high-speed steel.

As with ordinary high-speed steels, "Substitute 66" steel is annealed by heating through the upper change point and cooling slowly in the furnace. A temperature of 850°–875° C. will ensure that condition. Packing in air-tight boxes is the best treatment for tungsten high-speed steels, and is desirable for molybdenum types. The hardness after annealing should be approximately 220 Brinell, and the steel should definitely be easier to machine than its counterpart of the tungsten variety.

Hardening

The most important factor to be observed in hardening "Substitute 66" steel is that a lower hardening temperature must be used than is customary for other high-speed steels. Fig. 1 shows a typical hardening curve for "66" type molybdenum high-speed steel. The broad line shows variations in hardness which can be expected in actual use. It will be observed that the peak in the curve occurs at a hardening temperature of 1,200°–1,250° C. Above this temperature the hardness decreases owing to increase in the amount of austenite. (Incidentally, the diagram is for untempered specimens only.) It will be readily understood that hardening temperatures as low as 1,175° C. can be used where minimum grain growth—i.e., maximum toughness—is required.

For normal hardening a range of 1,200°–1,260° C. is available. If temperatures such as are common for 18% tungsten steels are used, brittleness will develop rapidly. As a general guide for practical purposes, it is suggested

that the hardening range given above should be divided into two ranges—1,200°–1,230° C.,—the lower range being used for tools requiring a measure of toughness, such as drills, countersinks, reamers, etc. For tools which are sufficiently robust to withstand a greater amount of shock, the hardening range 1,230°–1,260° C. will ensure increased hardness and better performance, but coarser grain size is probable. Lathe tools and planer tools are outstanding examples of tools which should be hardened in the upper range.

The well-known phenomenon of "sweating", which is probably due to fusion of the oxide formed on the surface of steels and which is prevalent in tungsten high-speed steels, does not occur in the molybdenum steels. Consequently, operators who use the old type of furnace with an oxidising atmosphere will not be able to observe the times at which the material under treatment should be removed. Sweating in ordinary tungsten high-speed steels is not observed in controlled atmosphere furnaces because no oxide is formed. The use of salt baths also entirely prevents its formation.

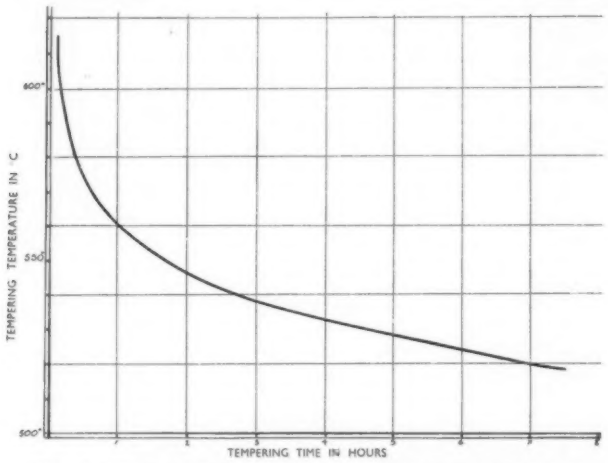


Fig. 3.—Curve showing time of maximum secondary hardness.

Hardening in these furnaces is done by the time and temperature method, the work being held at hardening heat for definite periods varying from a few seconds to several minutes. Hardening by "sweating" is old-fashioned, uneconomical and haphazard—the very appearance of the bubbles which occur with this method indicates that poor surface conditions will result.

There is no difficulty in arranging a programme by the time-temperature method, which is commonly used in large-scale production of taps, drills, reamers, hacksaws, etc. Fracture pieces of the same size and shape to be treated are hardened by holding at the selected temperature for increasing periods of time. The hardness of each test-piece is taken and each piece broken. It is thus easy to determine a hardening time which will not give a coarse fracture, but will give adequate hardness.

It should be remembered that for a selected hardening temperature, the hardening time will depend on the size of the piece to be treated, and this detail should be calculated for every size of article and used consistently. Various types of clocks can be obtained for these timing operations. The times of hardening will, of course, depend upon the thermal capacity of the furnaces; times which are correct for one furnace are not necessarily suitable for another furnace. Salt baths, because of their rapid rate of heating, will require short heating times.

Pre-heating for molybdenum steels follows the same procedure as for tungsten high-speed steels, and should be done at a temperature of 800°–850° C. In good hardening practice for bulk production this operation is timed. The time of pre-heating should not be so long as to cause scaling, but should be adequate to give even heating of the work.

Where some guidance is required for hardening, it should be appreciated that at the maximum temperature of 1,260° C. molybdenum high-speed steels will coarsen more quickly than the tungsten type, so that less hardening time can be allowed. At the minimum temperature of 1,200° C. the rate of coarsening is not so rapid and actually a longer hardening time than for 18% tungsten steel can be used.

Quenching

Oil quenching or air and salt quenches may be used for molybdenum high-speed steels, subject to the usual variations necessary for complicated shapes. Many users of high-speed steels do not appreciate that after the steel is quenched, hardening does not actually take place until a low temperature is reached. Until this change occurs the steel is comparatively plastic, and straightening operations are carried out much more easily at lower temperatures when the internal metallurgical change has taken place. Reheating after the steel has been allowed to go cold does not produce the same conditions.

Tempering

Fig. 2 shows the effect of tempering on the hardness of "Substitute 66" steel when hardened at two different temperatures and subsequently tempered. It will be seen

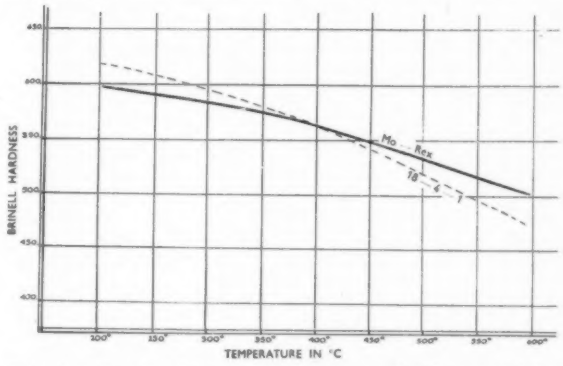


Fig. 4.—Hot hardness of 6/6 type molybdenum high-speed steel.

that the general outline of the curve is the same as that for tungsten high-speed steel. From the hardened condition there is a gradual fall in hardness, followed by an increase to a point at which a rapid fall commences as the temperature increases. The peak of this curve, usually known as "secondary hardening" occurs, it will be noted, at a lower temperature than for tungsten high-speed steel, at 550° C. instead of at about 600° C. It follows, therefore, that the tempering of molybdenum high-speed steel is carried out correctly at about 550°–580° C.

The greatest care should be taken if, to reduce the hardness, temperatures higher than this are used, because of the marked rapidity with which hardness falls. Lower hardness values are achieved much more easily by the use of lower hardening temperatures.

Time of Tempering

American investigators[4] state that a longer time is necessary for the tempering of molybdenum high-speed steels than for the tungsten varieties, because of the persistence of the molybdenum austenite. Fig. 3 illustrates this by showing the times necessary to give maximum secondary hardness at various tempering temperatures. The line of the curve shows a time of approximately 1¼ hours at 550° C., and this compares with results obtained in actual practice of 1 hour minimum and up to 2 hours. The use of a 1-hour minimum period for tempering molybdenum high-speed steels is advised. Ordinary tungsten steels can be tempered satisfactorily in about 30 mins.

4 J. P. Gill and R. S. Rose. "Metal Progress." September, 1941.

Red Hardness of Molybdenum High-speed Steels

Some investigators claim that the hardness of high-speed steel at high temperatures is the true test of its resistance to wear in cutting. If these opinions are accepted, "Substitute 66" steel should be superior to 18% tungsten high-speed steel. The curves in Fig. 4 show the Brinell hardnesses and the results were obtained by the impact method, using a 10 mm. ball and a comparator specimen of known hardness on top of the ball, the specimen being held in a small furnace on an anvil block. A force is used which will give a reasonably sized impression according to the thickness and size of the specimen. It is not necessary to know the force of impact.

It is hoped that these views and comments will be of help and guidance to the user of alternative high-speed steel. The historical references will, it is hoped, give some idea of the years of work which were necessary to arrive at the present stage of development, and will engender a greater measure of confidence in all potential users.

The Significance of Hydrogen in Steel and Cast Iron

By D. W. Rudorff, A.Am.I.E.E. M.Inst.F.

Results of investigations on the influence of oxygen and hydrogen on the properties of iron and steel indicate that under certain conditions their effect is, in some respects, comparable with the effect of certain metallic alloying elements. Some of these investigations are reviewed, and attention is directed more particularly to the influence of hydrogen on iron and steel.

INCREASED attention has been paid in the last twenty years to the influence of oxygen and nitrogen content upon the properties of iron and steel. In the case of ordinary carbon steels, research regarding the effect of nitrogen was not so much concerned with the influence of this component upon the hardenability of the steel as with its effect upon ageing; while in the field of alloy steels, a considerable influence of nitrogen content upon hardenability and austenite stabilisation could be observed. These results led to the employment of nitrogen in corrosion-resistant alloys in place of nickel, for the purpose of augmenting austenite stability.[1] Additions of nitrogen are also used in cementation processes, with a view to increasing hardenability of the surface stratum.[2]

Regarding oxygen, the fundamental investigations of H. W. Quaid and E. W. Ehn[3], concerned with abnormal steels, led to the conclusion that the oxygen component in alloy steels has an adverse influence upon the hardenability of iron and steel alloys. While subsequent research carried out by F. Duftschmid and E. Houdremont[4] has cast doubt upon the correctness of this conclusion, a secondary influence of oxygen, brought about by the oxidation of certain components in alloy steels, in some cases may serve to stimulate a diminution in hardenability. Similarly, non-metallic inclusions may facilitate and lead to transformation, thus tending towards a decrease in hardenability.

American opinion tends to regard abnormality as a property of pure iron, although there are still quite a few adherents of the oxygen theory. One of the facts usually cited as proof that oxygen is the cause of abnormality is that in the cementation of ordinary carbon steels, in certain cases either a normal or abnormal structure can be obtained according to the cementation process chosen. In cases of this kind, with the employment of cementation powders (in which the cementation effect is chiefly due to the action of the oxygen-containing carbon-monoxide) the steel may become abnormal; while cementation with town's gas leads to a steel with normal structure. In the case of cementing powders, oxygen-action was assumed to be responsible for the creation of the abnormal structure. Houdremont reports[5] that on the occasion of a discussion of this matter with American metallurgists, the latter voiced the opinion that the tendency of pure iron towards abnormality must not *per se* be accepted as proof of the invalidity of the oxygen theory. Even the most minute oxygen content in the iron might be quite sufficient to create abnormality.

Should this latter indeed be true, it was argued, similar effects might be caused by other elements contained in the iron in smallest quantities, such as, for instance, hydrogen. This argument has now been substantiated by the series of tests conducted by Houdemont and Heller.[5] The tests of these investigators were carried out with pieces of 30-mm. diameter, made with eutectoid ordinary carbon steel. The test procedure consisted in heating the pieces to 1,000° to 1,100° C. in a dry hydrogen atmosphere for 5 to 10 hours. For the sake of comparison, identical pieces of the same material, packed in burnt-out charcoal, were subjected to the same thermal treatment in ordinary atmosphere so as to obtain identically coarse structure.

In order to prevent the escape of hydrogen during the cooling period, the test-pieces were quenched in water from the heating temperature. Micrographic inspection showed that the hardening depth was increased by the hydrogen absorption. This effect was also found to be somewhat increased when the time of treatment was raised from 5 to 10 hours. Both the hydrogen treated specimens and the charcoal-packed samples showed identical coarseness of grain.

Referring to Fig. 1, which illustrates the results obtained with 1,000° C., it is seen that at this temperature the hydrogen treatment leads only to a diminution of the extent of the soft core; while the corresponding test run made with hydrogen at 1,100° C., shown in the right-hand graph of Fig. 2, resulted in an almost complete hardening of the entire cross-section of the piece. The influence of hydrogen upon the depth of hardening is thus clearly demonstrated.

It might be argued that this increase in hardenability is not caused by the hydrogen itself, but is rather due to attendant conditions such as reduction of oxygen, or disintegration of sulphates and their removal. In order to disprove this line of argument Houdremont and Heller decided to expel the hydrogen, by heating, from the pieces investigated, and to harden them again after increasing their grain fineness by normalising. It was found that pieces treated in this manner exhibited identical depth of hardening and sensitivity to overheating, as was found in the pieces not treated with hydrogen. The result of this investigation proved beyond doubt that the increase in hardenability was due solely to the action of the hydrogen.

In the case of alloyed steels, it was found that with completely transforming steels the action of hydrogen

1 W. Tofaute and H. Schottky, *Arch. Eisen H. Wesen*, Vol. 14, 1940-41, p. 71.
2 R. J. Cowan and J. T. Bryce, Trans. Am. Soc. Metals, Vol. 26, 1938, p. 766.
3 H. W. Quaid and E. W. Ehn, Trans. Am. Inst. Min. Met. Engrs., Vol. 67, 1922, p. 341.
4 F. Duftschmid and E. Houdremont, *Stahl und Eisen*, Vol. 51, 1931, p. 1,613.
5 E. Houdremont and P. A. Heller, *Stahl und Eisen*, Vol. 61, 1941, p. 756 *et seq.*

leads to a delay in transformation. It must, however, be noted that the hydrogen-treated pieces showed a somewhat coarser grain. But this alone cannot explain the pronounced differences in the transformation characteristics. It must, therefore, be assumed that in view of the influence of the cementation process upon the structure, oxygen cannot be held responsible for the formation of abnormal structures One may even assume that the formation of the normal structure is actually due to the action of the hydrogen in town's gas cementation. This action may be analogous to that of other elements causing increased hardenability, as, for instance, manganese or silicon.

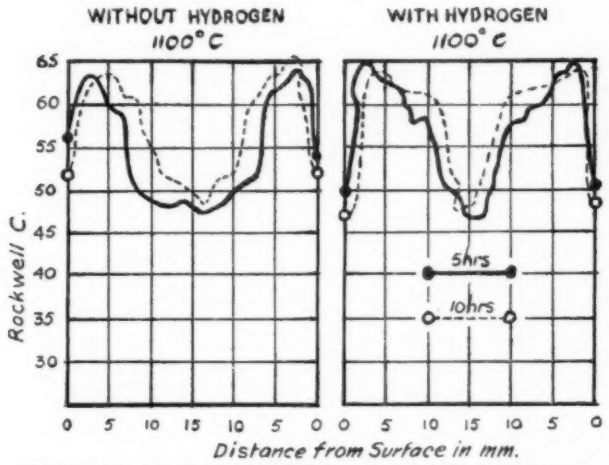


Fig. 1.—Increase in hardenability owing to hydrogen absorption. Steel composition 0.46% C, 0.13% Si, 0.18% Mn. Heating temperature, 1,000° C.

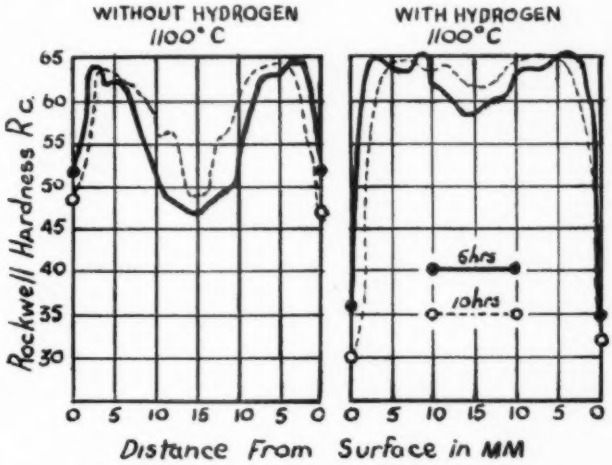


Fig. 2.—Influence of hydrogen upon depth of hardening. Steel composition, 0.46% C, 0.13% Si, 0.18% Mn.

In order to investigate the problem still further, two test-pieces of mild steels of different composition were subjected to tests designed to show the influence of hydrogen upon the structure formation in the cementation process. The two steels were so chosen that their cementation with town's gas at 920° C. yielded an abnormal structure. One of the steels selected was a carbonyl steel containing 0·01% C, 0·01% Si, and 0·03% Mn ; while the other steel was a material containing 0·04% C, 0·01% Si, and 0·11% Mn. Of these, the carbonyl steel was more pronouncedly abnormal than the other material, the latter showing signs of abnormality only in spots. Test specimens of these two materials were heated for 10 hours at 1,100° C. in hydrogen and in nitrogen atmospheres. Immediately subsequent to this treatment, the test-pieces were placed in a heated container for town's gas cementation.

After hydrogen treatment, the carbonyl steel showed somewhat greater pearlite formation than after nitrogen treatment. The other steel was found to have become completely normal after hydrogen treatment, but pieces of this material subjected to nitrogen treatment were found to be markedly normal. Both steels had exhibited identical MacQuaid-Ehn grain size before cementation.

According to Houdrement and Heller, these tests fully justify the conclusions drawn with regard to the influence of hydrogen upon structure formation in the cementation process. The increase in hardenability of tool steel corresponds to a delay of the transformation process under the influence of the hydrogen absorption. Investigations regarding the influence of hydrogen upon the red-hardness brought negative results. This was, of course, to be expected, as it is well known that at 200° C. the hydrogen is lost from the material.

With regard to recent investigations into flake formation,[6] it is important to note that the hardened hydrogen-treated pieces exhibited fairly large flake-like cracks as well as microcracks. Contrary to the experiences related by another investigator,[7] these microcracks could not be made to disappear by annealing.

The conclusions drawn with regard to the effect of hydrogen upon steel were entirely confirmed by observations made with cast iron. The influence of hydrogen upon cast iron was recently investigated by A. Boyle,[8] according to whom hydrogen favours the formation of primary carbide and also of eutectic carbide, during the solidification of cast iron. In view of the fact that no statement had been made by Boyle regarding the secondary cementite-decomposition, special attention was given by Houdremont and Heller t this matter, which is particularly significant in connection with the effect of the hydrogen component. The experiments made by Houdremont and Heller were carried out with machined test-pieces of cast iron containing 2·86% C, 2·21% Si, 0·73% Mn, 0·37% P and 0·07% S. In its original state this material showed an exclusively pearlitic structure.

Test-pieces of this material, measuring 15 mm. in diameter and 100 mm. in length, were heated at 1,000° C. for 2, 4, 6 and 8 hours in dried hydrogen containing 0·2 g. H_2O per cubic metre, and in moist hydrogen containing 0·77 g. H_2O per cubic metre, and were subsequently cooled in the furnace to room temperature by subjecting them to a stream of hydrogen.

Other pieces were held at 1,000° C. for 6 hours in ordinary furnace atmosphere. As was to be expected, the slow cooling, through critical point A_1, led to a practically complete ferritisation of the ground mass, the amount of combined carbon amounting to no more than 0·15%. In the case of the pieces exposed to the dry hydrogen atmosphere, a distinctly different result was obtained. Here, heating and subsequent slow cooling, through A_1, remained ineffective, and the pearlitic structure remained unchanged. Indications are that increased length of hydrogen exposure even tends to decrease the ferrite portion. In the pieces treated in the moist hydrogen atmosphere, the portion of pearlite disintegrated into ferrite and graphite was found to increase steadily with increased length of treatment, without, however, equalling the results obtained with heating in the ordinary furnace atmosphere.

While these tests serve as a good illustration of the effects obtained with heating in hydrogen, they do not yet permit of the drawing of an unequivocal conclusion as to whether oxygen or hydrogen is the determining factor. The circumstance that the effect of the moist hydrogen atmosphere is similar to that of the ordinary furnace atmosphere, would seem to suggest that the decomposition

6 H. Bennek and G. Klotzbach, *Stahl und Eisen*, Vol. 61, 1941, p. 597, 624. E. Houdremont and H. Schrader, *Stahl und Eisen*, Vol. 61, 1941, p. 649, 671.

7 E. C. Bain, *Journ. Iron and Steel Inst.*, Vol. 138, 1938, p. 33.

8 A. Boyle, Am. Inst. Min. Met. Engrs., Techn. Publ. No. 809, p. 1,046; also H. A. Schwartz, G. M. Guiler and M. K. Barnett, Trans. Am. Soc. Metals, Vol. 28, 1940, p. 811.

was accentuated by the diffusion of oxygen into the material. One might also assume that the hydrogen treatment leads to the removal of oxygen in the material; that is to say, that it is the removal of oxides, and not the enrichment with hydrogen, which must be held responsible. This contention sounds rather unconvincing, particularly in view of the fact that according to the statements of other investigators,[9] oxygen rather tends to increase carbide stability.

In order fully to clarify this matter the following series of tests was run by Houdremont and Heller, in which samples of identical size were held at 1,000° C., with subsequent cooling in the furnace at a speed of approximately 4° C. per minute, to wit:—

(1) Three hours at 1,000° C. in oxygen-free argon, of 0·0554 g. per cubic metre moisture content.

(2) Six hours at 1,000° C. in oxygen-free nitrogen, of 0·0987 g. per cubic metre moisture content.

(3) Six hours at 1,000° C. in oxygen-free hydrogen, of 0·114 g. per cubic metre moisture content.

Because of the shortness of test (1), caused by lack of sufficient argon for a six-hours test, absolutely complete decomposition into ferrite and graphite could not be achieved. But it could be obtained in test run (2). Test (3), however, failed to reveal any considerable changes in structure at all. It is thus quite clear that graphite decomposition could not have been initiated by oxygen penetration.

In order to ascertain also whether it is the oxygen contained in cast iron, or whether it is the action of the hydrogen upon oxides, which is of decisive influence, further tests were made. These were based on the assumption that the process must be reversible if hydrogen is the responsible agent. Thus, it must be possible to produce anew a pearlitic structure by re-treating the ferritic piece in a hydrogen atmosphere at a temperature exceeding A_1. For this purpose a piece was held for 6 hours at 1,000° C. in a normal atmosphere, and subsequently at the same temperature and for the same length of time in a hydrogen atmosphere. Examination showed that the ferritic structure was indeed transformed back into a pearlitic one. In order to ascertain whether this re-transformation was really due to the direct influence of the hydrogen, the pearlitic sample was then held for 6 hours at 1,000° C. under a vacuum of about 0·2 mm. Hg: this was followed by slow cooling.

Subsequent inspection of the piece showed that the removal of hydrogen had resulted in the renewed decomposition into ferrite and graphite. Degasification of a number of specimens by heating at 900° C. yielded an average hydrogen content of 0·71 cub. cm. per 100 grms.

Piece No.	Treatment.	Vacuum mm. Hg.	H_2 cub. cm./ 100 g.	Co cub. cm./ 100 g.	Structure.
0	Not treated.	—	—	—	Pearlitic.
1	Degassed for 3 hours at 900° C.	0·005	0·5	0·72	Ferritic.
2		0·004	0·6	0·39	,,
3		0·0025	0·56	0·32	,,
4		0·002	0·66	0·4	,,
5	Degassed for 3 hours at 900° C., thereupon enriched with hydrogen for 15 hours at 900° C.; cooled in furnace.	0·005	0·87	0·32	Pearlitic.
6	Degassed for 3 hours at 900° C., thereupon enriched with hydrogen at 900° C.; cooled in furnace, and	0·0001	1·1	—	—
	Degassed for 3 hours at 900° C.	0·0001	0·78	—	Ferritic.
7	Enriched with hydrogen for 17 hours at 900° C., water-quenched; subsequently degassed for 7 hours at 900° C.	0·001	3·45	—	Ferritic.

By applying hydrogen treatment to the degassed pieces without previous cooling, the original hydrogen content could be fully restored.

A further series of tests was made at 900° C. in order to ascertain the magnitude of hydrogen absorption at this temperature. As the subjoined table shows, an average hydrogen absorption of 3·45 cub. cm. per 100 grms. was obtained. But this figure does not take into account the quantities of hydrogen escaping during quenching and handling of the pieces. Complete disintegration into ferrite and graphite after degassing was not, however, obtained at this temperature.

The results of these investigations were summarised by Houdremont and Heller to the effect that in the case of cast iron, hydrogen can be considered to play the role of an alloying element acting towards carbide stablisation, the required hydrogen content being 0·5 to 0·7 cub. cm. per 100 g. No explanation of the nature of this action can be offered as yet. The hydrogen-enriched pieces show well-formed lameller pearlitic structure and graphite. Hydrogen must, therefore, be considered to be somehow instrumental in causing structure formation to proceed towards the metastable iron-cementite system. In the case of steels tending to exhibit abnormal structure formation, hydrogen enrichment can be used to effect a normal cementite and pearlite formation. In certain respects the action of hydrogen is, therefore, comparable to that of manganese and chromium.

9 H. Jungbluth and H. Bruggs, *Techn. Mitt. Krupp, A. Forsch. Ber.* Vol. 1, 1938, p. 171.

The Oxidation of Cerous Sulphate at a Rotating Anode

THE significance of motion of an electrode surface, relative to the electrolyte, on the reaction occurring during electrolysis, has received considerable attention during the last decade. The effect of relative cathode-catholyte motion during electrodeposition of metals has been studied with special care, and a number of theories have been proposed to account for the effect of such motion on equilibrium electrode potentials in non-electrolytic cells. No efforts, however, seem to have been directed to a study of an anodic oxidation under such conditions, but the results of a preliminary study of a system of this type is given in a recent paper by J. L. Culbertson and C. Rutkowski, which deals with the effect of anode rotation on the current efficiency of the electrolytic oxidation of cerous sulphate.

In this study the authors noted that the anode rotation causes an increase in the current efficiency and that the mathematical form of the relation of the latter quantity to the rotation speed is similar to that of a heat transfer coefficient of a liquid film, at a solid-liquid interface, to the speed of the fluid past the solid surface. Since the heat transfer coefficient is diffusion controlled and its magnitude is increased by film attrition, it is concluded that the electrolytic reaction is governed in a similar manner.

Inspection of Arc-welded work

A MEMORANDUM has recently been issued by the Advisory Service on Welding, through the Ministry of Supply. It is the seventh of a series which deal with various aspects of welding as applied to Government work. This publication is concerned with the inspection of arc-welded work and is designed to facilitate the work. The inspection procedure recommended provides a list of points to be examined and a common basis of comparison for different inspectors. The procedure given does not apply to the inspection of welded armour plate.

The principal factors affecting welded work are: the weldability of the base metal; the arc welding machines and equipment; the electrodes; the welding procedure; the control of distortion; and heat treatment; and the notes given in this memorandum provide a system of inspection which will assist in maintaining the required standard of welding.

The memorandum is distributed free to firms using welding equipment and firms requiring additional copies for distribution to their design, production and inspection staffs and to welding operators, chargemen and foremen may obtain extra copies from the Advisory Service on Welding, Department of the Controller-General of Research and Development, Ministry of Supply, Great Westminster House, Horseferry Road, London, S.W. 1.

Tin Patents and the War

By S. T. Madeley

The position with regard to patents is not too clear, and the author suggests a thorough overhaul of our whole patent system; meanwhile, inventors should continue to patent their inventions in order to assist the commercial prosperity of the nation.

THE turn of events in the Malay Peninsula has drawn attention to the importance of tin in the present world war. Those engaged in the industries connected with tin, whether it be tin alloys or tinplate, or tin pigments, must inevitably be considering how best to make use of their present and future supplies of this important metal. For although the United States, which drew much of its tin from Malaya will, more immediately than us, feel the restriction of its source of supply, we must not forget that much of our war material comes from U.S.A., though our tin itself comes from Bolivia and Nigeria.

New and improved methods of extraction and recovery of tin will have to be evolved. This means "invention." No doubt in the United Kingdom our efforts in this direction will chiefly be directed towards recovery.

Inventions are industrially handled through patents. It will be of interest, therefore, to examine the graph of patents relating to tin over the last thirty years. :—

TABLE SHOWING APPROXIMATE ANNUAL NUMBERS, AS ESTIMATED BY THE WRITER, OF PATENTS GRANTED RELATING TO ALLOYS CONTAINING TIN, FOR THE PERIOD 1910–1941.

Year.	Patents.	Year.	Patents.	Year.	Patents.	Year.	Patents.
1910 ..	13	1920 ..	8	1930 ..	20	1940 ..	33
1911 ..	15	1921 ..	11	1931 ..	27	1941 ..	38
1912 ..	16	1922 ..	10	1932 ..	34		
1913 ..	14	1923 ..	13	1933 ..	20		
1914 ..	6	1924 ..	12	1934 ..	22		
1915 ..	7	1925 ..	15	1935 ..	20		
1916 ..	5	1926 ..	15	1936 ..	31		
1917 ..	5	1927 ..	16	1937 ..	43		
1918 ..	6	1928 ..	16	1938 ..	32		
1919 ..	8	1929 ..	17	1939 ..	44		

The above data are illustrated in the accompanying graph.

An examination of the curve first draws our attention to the steady rise therein since the last war. A very respectable peak was reached in 1932. This agrees substantially with a peak period in the patenting of cans, canisters, and so on, employing tinplate. The subsequent fall in the curve also coincides with a drop in the inventions, of course. The recovery in 1937 and subsequent thereto of tin alloy patents is considerably larger than that which occurred as regards the other branch of the industry above mentioned.

There is no doubt that the war period recovery of tin-alloy patents is connected with the present increased demand for tin. Some 50% of our tin finds its way into alloys. Let us briefly glance at a few British, Canadian and U.S.A. Patents. This will give us an idea of the lines of recent invention in connection with our subject.

British 534,797 (Malloy Metallurgical Products Ltd.).—An electrical make-and-break contact. The contact consists of an alloy as follows : 1 to 15% boron ; 0·5 to 3% tin (or cadmium, copper, gold, magnesium, manganese, palladium, platinum, zinc) ; balance, silver.

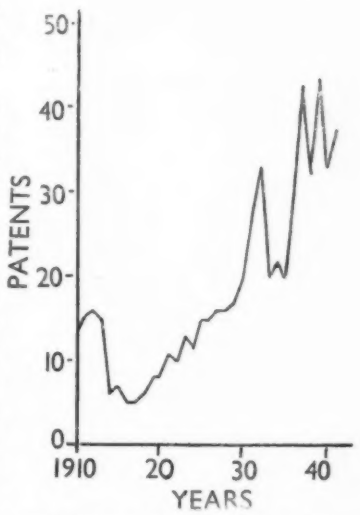
British 523,303 (National Lead Corporation).—Removal of aluminium fron tin or alloys containing tin—for example, alloys of lead and tin, where the aluminium is present as an impurity. The tin or the alloy containing tin is melted and treated with alkali metal carbonate in the molten state to remove the aluminium.

Canadian 398,655 (Cerrode Pasco Copper Corporation).—Separation of tin from impure lead. Into the molten bath of lead oxide at about 950° F. is stirred a disintegrated stream of lead oxide and air. The stirring is performed by a disintegrating impeller immersed in the bath at a speed of 900 r.p.m. or over, so as to form a vortex extending to the impeller. Lead oxide and air pass into the vortex and into contact with the impeller. The disintegrated stream is projected into the bath and mixed therewith.

United States 2250843 (Robinson).—Production of tin. A solution of iron and tin in hydrochloric acid is electrolysed. The iron is retained in the ferrous condition by the tin being kept substantially wholly in the stannic condition during the electrolysis.

United States No. 2250552 (Pinter).—Production of pure stannic oxide. Metallic tin in a reaction chamber is converted without residue into tin sulphide vapour by reacting on pure metallic tin with sulphur vapour, the sulphide vapour being maintained in excess of that reacting with the tin. Substantially pure stannic oxide is obtained from the tin sulphide vapour by oxidising the vapour in a separate reaction chamber.

Our next step is to consider how the Patent Acts affect the situation.



Prior to the last world war, the Patents, etc., Act contained no provisions, we may note, whereby a patentee could obtain relief solely on the ground of loss or damage due to hostilities. As patents then had a normal term of only 14 years, a loss of four, five or six years due to dislocation of business during war-time might mean irreparable loss to the patentee, for it is well known that in the ordinary way the last few years of a patent's life are the most valuable.

Measures, therefore, had to be improvised to meet the situation over 1914-1918. The Comptroller-General granted emergency licences under enemy patents. Section 309 of the Peace Treaty safeguarded the holders of such licences against post-war infringement actions. Sections 307, 308 dealt with the rights of patent applications made under the International Convention etc. The belligerent countries granted a moratorium.

The amended Patents Act of 1919 included Section 18 (6), under which a patentee could apply for an extension of term of his patent solely on the ground of loss or damage due to war or hostilities. In addition, the patent term was extended from 14 to 16 years.

Another most important event was the setting up of a Royal Commission for awards to inventors. The Commission could grant up to £50,000 in respect of a patented invention and recommend the Treasury to grant more. Unpatented inventions were left to the consideration of the Treasury. Many large awards were made. There was also a general moratorium.

Let us now look at the present position. The damage caused by air raids to factories and manufacturing plant generally, both owing to actual destruction and to loss owing to the need for decentralisation and so on, is further

accentuated by the depletion of personnel through national service in one form or another. This will undoubtedly cause considerable loss to patentees in many cases, as the life of a patent will practically be shortened by years—perhaps several years. If this happens during the maturity of a patent—namely, towards the end of its life, when in the ordinary way the profits should mainly accrue—the matter may be very serious. Supposing that the final result of hostilities is to "dock" the patent of five years of life, and assuming, for example, that the patent for, say, a tin alloy was calculated to bring in some £83,000, then there may be a loss of, say, £28,000 or £29,000. It is here that Section 18 (6) may be of great assistance provided the case is wisely handled. Patentees should collect their evidence in good time and whilst still easily available, and seek professional advice as soon as may be. So much for Section 18 (6).

We will now turn to the Patents, etc. (Emergency) Act, 1939, which came into force on the outbreak of war, and was moulded on the experience gained in 1914-1918. Similar provisions came into force in Canada.

The main section in this Set, which concerns patentees, are numbers one, two, four and six. They give the Comptroller-General of Patents certain valuable and important powers. By Section 1 he is enabled to revoke or vary an existing licence granted under a patent, say, for tin-plating owned by an enemy or enemy subject, on application by a suitable party. Section 2 gives the Comptroller power to grant an emergency licence under a patent for, say, recovery of tin, on similar conditions. The licence will be so drawn as to protect the licensee against a post-war infringement action intended to deprive him of the benefits of any business which he may have built up under the licence during hostilities. Under Section 4, a patent for, say, a tin-plating process can be granted to an enemy, but the enemy will not be able to require delivery of the patent. Usually, however, the application will only be carried to the acceptance stage unless somebody applies for a licence under Section 2. This Section eaves the International Convention in force.

Very helpful provisions are to be found in Section 6. According to this Section, the Comptroller may, if he is satisfied on the evidence, grant extensions of time for paying patent fees, filing patent documents, and so on, where delay is due to the war, and enforcement of the period allowed in the ordinary way would be injurious to the party concerned.

A Bill embodying a new Patent Act is before Parliament at the time of writing and will no doubt shortly be coming into force. It is not anticipated, however, that it will alter the state of affairs much. What is required, of course, after the war, is a thorough overhaul of our whole patent system, directed to the encouragement of invention, if we are to survive as a mercantile nation under the stress of post-war years.

May the writer indicate a few points which, in his opinion, are desirable?

There should be a Royal Commission for awards to inventors permanently sitting, which should have power to make awards irrespective of evidence of commercial use or success. Often a pioneer reaps practically nothing but disappointment, and mere followers secure all the advantages.

Two years should be allowed between filing a provisional specification and the completion of the specification.

The Convention and like periods should be extended to two years.

"Communicated" patent applications should either be dispensed with or strictly controlled.

All patents should be immediately extended for two years, thus increasing the term to 18 years.

Patents legal procedure should be cheapened.

Greater latitude should be exercised in construing basic patent documents in Convention cases.

Our Patent Office should be enlarged, and the prospects of the staff improved.

Meanwhile, we should continue to patent our inventions here and abroad, as on them depends, or will depend, in large part, our commercial prosperity and necessary export trade.

Vickers Ltd.

Vickers, Ltd., announce that the net profit for the year 1941 is £908,905, against £1,119,120 in 1940. The transfer to contingencies is the same as in 1940—viz., £250,000—and the carry forward £690,786, after paying a final dividend on the ordinary stock of 6% actual, less income-tax, making 10% for the year ended December 31, 1941.

The Hardening throughout of Chrome-Molybdenum and Chrome-Nickel Case Hardening Steels*

THIS is an extension of a former investigation which referred to heat-treated chrome-molybdenum steels only,† while Pomp and Krisch deal with casehardening steels, and try to find out what qualities can be expected in the core of such steels of diameters of 0·2 in. to 2·4 in. (5 mm. to 60 mm.).

Rods of two chrome-molybdenum, one chrome-nickel-molybdenum, and two chrome-nickel steels were tested by measuring the hardness at the surface and over the section, as well as by tensile and notch-bar impact tests on specimens taken from the surface and from the core of the rods. It was not possible to obtain the figures of strength, as specified by the German Air Ministry, for all dimensions of the same material with equal heat-treatment. They were, as a rule, higher with the thin and lower with the thick dimensions. The notch-bar impact tests did not show brittle fractures in any case. The ratio 0·35 was determined for tensile strength to hardness in the cross-section.

The Spectrum-Analytical Determination of Elements in Electric Sparks by Direct Photo-Electric Measurements of the Intensity of the Lines ‡

IN order to directly determine elements in electric sparks by the spectrum-analytical method, Thanheiser and Heyes divided the light of the sparks into two parts, and each part was analysed separately. One line of each part was cut out and the light of these two lines were brought into contact separately, each with a photo-cell. The two photo-currents of these cells have been measured by two observers, and the ratio of the result has been determined during a given time. The values of these quotients are proportional to the ratio of the intensity of the lines. Therefore, as the connection between the intensity of the lines and the concentration of the element in the specimen is known, it is possible to determine the latter.

The mean error of measurement when using this new method has been found to 1·49%. It has been applied for determining manganese and magnesium in aluminium, and has been proved quite useful. A comparison with the hitherto usual method of photographic measurement shows that both give nearly the same accuracy of analysis.

* Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung zu Düsseldorf. Vol. XXI. No. 21.
† *Metallurgia*, October, 1939, p. 196.
‡ Mitteilungen aus dem Kaiser-Wilhem-Institut für Eisenforschung zu Düsseldorf. Vol. XXI. No. 22.

Metallurgical Furnaces and the Mechanical Stoker

By Glyn Powis

Although mechanical stokers have long been applied in steam raising, their application to metallurgical furnaces is comparatively recent. Several installations for heat furnaces of different types are operating with complete satisfaction, and it seems that for certain forms of heating they offer several advantages. Their application is discussed.

FOR many years prior to the war manufacturers were exploring every likely avenue in the hope of reducing their heating costs without the sacrifice of efficiency. The problem has been solved to some extent, and not the least important development has been that of the mechanical stoker. Developed originally for boiler work of marine and central-heating type, this type of machine has been applied successfully to the heating of metallurgical furnaces. One important feature has been that whereas in certain instances, notably drop-forging, coal-firing by hand failed to provide suitable temperatures, the even feed of the stoker has remedied the deficiency. The method of feeding the coal is by worm conveyer, which has been proved to be both satisfactory and economical in that the coal is gradually pushed into the hot bed of fuel above the tuyeres. The gradual movement is responsible for a steady rise in temperature of the fuel, causing the liberation of volatile and smoke-forming gases which become burnt completely by the incandescent part of the fire above. This transforms the original coal into coke, which also burns on top of the fire bed. Thus the heat contents of both gas and coke formed from the coal are used to the maximum.

Advantages Offered

Some of the advantages offered by the use of stokers include the saving in electrical power, since only one-quarter to one-fifth of the energy is necessary to supply air blast, as compared with that for gas or oil. They give individual control, therefore the cost of running is always proportional to the number of fires working. The closing down of one or more furnaces feeding a battery of stamps always represents a saving. Most other fuels draw their air supply from a common main, and reduced fire numbers result in higher operating costs for those remaining. The fuel costs are greatly reduced, as bean-size nuts are useable, which are available at a low cost, compared with oil.

A mechanical stoker of this type working correctly gives off less smoke than the average oil fire, which effects a considerable saving in painting and window cleaning. A more important advantage, however, is the condition of the metal when heated; it is of a mellower nature than from an oil fire, the temperature being more even, providing what is known as a "soft, wet heat." The "soft" heats are usually the most suitable for stamp and press work, the metal flowing evenly in the die-impressions. Further, less slag is found in the stoker-fed fires as the "heats" do not tend to drip on the edges as in oil fires. It is noteworthy also that considerable saving is effected in fire-building and repair costs. Arches and cramps, if built of a good siliceous (not a full silica) brick will last much longer than when using gas or oil.

The main disadvantages are the removal of ash and clinker from the combustion chamber and the filling of the hoppers with raw fuel. As most oil fires require slagging, however, the task of removing clinker from a stoker fire is cancelled out, leaving only the hopper filling outstanding. Gas shows up to advantage in this respect. In view of the great savings possible, however, manufacturers quickly overcome these small difficulties.

Furnace Design

With stoker usage for forge-fire application in its infancy, the question of furnace design is as yet largely undeveloped. Many manufacturers have bought machines with the intention of carrying out their own experiments, the stoker suppliers being in some cases quite as much in the dark as the potential user. Each new sale, however, adds to the experience of the seller, and that position will not obtain for long.

Up to the present there are two main types of fire. One type is the direct-heating or box type fire, the other is the reverberatory one, where the combustion chamber is built separately from the heating chamber, the flames being "pulled" through the latter by means of a small, damper-controlled stack. Each type has a distinct sphere of usefulness.

Box Type Fires

In the case of the box type fire, the dimensions of the actual structure are more or less pre-determined by the size of the stoker retort. Various capacity machines have different size retorts, and it is necessary to leave a "collar" around the tuyeres to accommodate ash and clinker. This "collar" may vary from 2 in. to 9 in. in width. Nothing less than 2 in. can be recommended, as this renders the tuyere sections liable to undue burning owing to the nearness of the clinker. Beyond 9 in. the fuel bed is apt to cover too large an area, and efficiency is reduced thereby. An approximately square hole of, say, 12 in. × 12 in. needs to be left in either one side or end of the structure to facilitate removal of the ash, etc. This hole may be closed by either a door, or several loose fire-bricks, and luted with clay.

As regards the height of the fires, by placing the opening or slot at a height convenient for the operator of, roughly, 3 ft. to 3 ft. 6 in. from ground level, sufficient room is left beneath to allow for the gradual building up of the fire bed during the course of the working day. If two or more openings are required on each side, they should be spaced in such a manner that the lower, or lowest, as the case may be, is not likely to foul the bed in the interior, and that the top one is not unduly high for the heater. Add to this a 9-in. arch and the total height is easily arrived at.

Let us suppose that a stoker has a retort which measures 27 in. × 14 in. around the outside of the tuyeres. Around the retort may be left a clinker space of 6 in. This would give an internal grate area dimension of 39 in. × 26 in. Build around this space the conventional 9 in. wall, and a gross dimension of 57 in. × 44 in. is obtained. In general, with a 9 in. wall almost universal for forge fires, it only remains necessary, after choosing the most suitable clinker space, to add these two dimensions to that of the retort, and the gross dimensions of any box-type fire is automatically settled. Many fires for furnaces to supply heat for specific work are readily designed and constructed.

One particularly design of fire is suitable for heating work, where only a portion of the steel section requires to be raised to the stamping or forging temperatures, or where work is heated on the billet or bar, such as in nut and bolt manufacture and most types of drop-stamping or forging.

The type of fire allows of the work being pushed through the opening in the side walls, protruding into the actual combustion zone. Where it is necessary to heat work all over, such as for billet rolling and certain pressing operations, a separate heating chamber is built alongside the combustion chamber—the whole forming one unit.

Reverberatory Fires

One drawback which some users may encounter is the difference in temperature between the two ends of the heating chamber. With the heat source at one end only it is difficult to overcome this completely. By a suitable baffle, however, together with an arch gradually reducing in height to the outgoing end, much can be done to ameliorate the trouble.

With suitable combinations of stoker size and chambers, it is possible to melt steel at one place in the chamber, and while it is not intended to convey that this purpose is the forte of mechanical stokers, it does serve to show what excellent results it is possible to obtain and also the wide range of possibilities as yet unexplored.

Modifications

The type of heating chamber described offers many means of variation. For high temperatures the centre of the chamber may have a channel into which tap cinder can drain, being tapped off as convenient through the side or bottom. For low temperatures the bottom may be built solid right across. Where evenness in temperature is important, it can be arranged to remove the waste gases through ports in a false bottom. This provides a useful means of keeping the hot flames low, and so filling the chamber. In cases of this kind it should be noted that no baffle as such is left at the end of the heating area, the end wall runs right across at this point, thus forming a natural baffle.

Other Applications

It is found frequently that stokers can be fitted to existing furnaces with but little alterations. In some instances a chamber may be added for the retort to form a combustion zone. One annealing furnace for malleable iron at least has been converted to mechanical firing in the manner thus described, while in some cases stokers are applied to the combustion chambers of existing core ovens. In such instances only very slight alterations were necessary. Mechanical stokers also serve a battery of machines being used for firing large steel plate heating furnaces. Abrasive wheel bonding and drying furnaces, pottery and brick kilns, and rolling mill furnaces are all ideal applications where conversion to stoker-firing merits attention.

Choice of Stoker

In order to ensure reliable service from the machines, it is essential to choose a make which is soundly built on engineering lines, and with this end in view it is not an easy problem to choose a stoker from those available. The machines are made in sizes capable of suiting practically any requirement from 15 lb. to 1,000 lb. per hour. In making a choice, when the particular size has been decided upon, it should be borne in mind that certain considerations call for attention.

As a self-contained unit, mechanical stokers provide their own blast supply by means of a small low-pressure fan driven by the same motor which operates the feed-screw. Ample blast is usually available at a pressure of from 1½ in. to 4 in. water gauge. This supply is conveyed to the retort by means of an air duct. At the retort it may pass either into a brick-built chamber covered over with "dead-plates," or into a one-piece cast-iron box, known as a blast-box. Certain manufacturers favour the dead-plate types, others the blast-box, while a few supply both descriptions.

While the dead-plate method can be satisfactory for lower temperature work, the plates are apt to warp to a certain extent under conditions of industrial firing. Warped plates are ill-fitting, which becomes aggravated by the rough removal of clinker, with the result that spent fuel, etc., not only finds its way into the bricked air-chamber, but blast finds its way out. The final result of this is inefficient combustion, due to the unevenness of the working.

The blast-box, which is simplicity itself to instal, suffers none of these difficulties, and being bricked around just below the level of the tuyeres, is practically immune from any risk of warpage or dislodgment. Furthermore, it is cheaper to replace one or two refractories as required than introduce a new dead-plate.

Coming into contact with the hot bed of fuel, the tuyeres, in order to give satisfactory service, require to be made of heat-resisting alloy iron. Many stoker manufacturers are now using the heat-resisting cast irons developed by the British Cast Iron Research Association for this purpose. Then the coal feed-worm cannot be expected to give lengthy service when supplied in malleable iron; for best results a molybdenum alloy steel is recommended.

Probably the most important part of the whole unit is the gearbox, which calls for discrimination in selection. It is here that the widest differences are usually found. Those boxes which have a good broad-faced gearwheel, with the remaining components of equal quality, offer a better chance of long and trouble-free service. Multi-coal-feed speeds are essential. Some suppliers fit as many as five speeds in addition to neutral. For obvious reasons, motors should be of the totally enclosed type. It is best that the stoker should have a motor which is well up to its task, rather than one of lower power which may labour on top feeds.

The hoppers should be of simple design and of stouter gauge material than is standard with central heating equipment. Industrial plant is always called upon to withstand rough treatment, and for this reason 12, or even 10-gauge sheets are to be recommended for their construction. One at least of the foremost manufacturers of stokers is employing a copper-bearing steel, in order to combat corrosion set up by wet fuel. Size of hopper should be borne in mind, and if possible should be such that filling is not necessary oftener than twice per one shift.

One further point worthy of mention concerns the refractories used in the furnace building. For the most part, walls and bottoms may be built of ordinary good quality, well-burnt firebrick. Bricks of low fusibility should not be used, as these slag away and tend to bind the clinker in the ashes. It is an advantage to "wash" out the bottom surrounding the retort, and also a distance of, say, 12 in. up the walls with magnesite peas and/or rubble. This helps to prevent the bricks being attacked by slag. For the upper parts of furnaces, including arches, a good class siliceous (not a full silica) brick containing approximately 85% to 90% SiO_2 will give a good life.

"Tin and Its Uses" No. 12

RECENTLY published by the Tin Research Institute, this issue has an article on tin and the war in the Pacific, which states that the present annual supply of tin is about 78,000 tons against a pre-war consumption of 110,000 tons. War requirements and buying for private stocks have brought the total demand recently up to a rate equal to 175,000 tons per annum. These figures make it clear that restriction of consumption must be applied swiftly, but skilfully enough to avoid dislocation to industry. All stocks must be utilised with the utmost discretion, and the less necessary uses must be suspended in favour of those indispensable to the war effort.

Among several other interesting and informative articles is one dealing with improvements in the smoothness of tin coatings on tinplate; another article describes a novel use of tinplate in iron foundries. The issue maintains the high standard of these publications.